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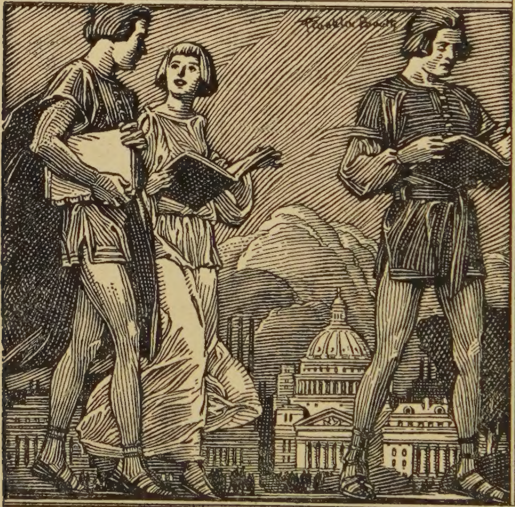

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STEEL AND ITS HEAT TREATMENT

VOLUME I. PRINCIPLES, PROCESSES, CONTROL

STEEL AND ITS HEAT TREATMENT

BY

D. K. Bullens

AND

The Metallurgical Staff of the
Battelle Memorial Institute

Fourth Edition

Vol. I. **Principles, Processes, Control.** The principles, processes, and fundamentals of control which relate to the heat treatment of steel in general, whether it be plain carbon or alloy steel. 445 pages, 6 by 9. 208 figures. Cloth.

Vol. II. **Engineering and Special-Purpose Steels.** A discussion of engineering carbon, simple alloy, complex alloy, and special-purpose steels. 491 pages, 6 by 9. 189 figures. Cloth.

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STEEL
AND ITS HEAT TREATMENT

Volume I—Principles • Processes • Control

BY

D. K. BULLENS

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THE METALLURGICAL STAFF

Of the Battelle Memorial Institute

FOURTH EDITION

SIXTH PRINTING

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FOURTH EDITION

Sixth Printing, August, 1944

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IN MEMORY OF

Albert Nelson Bullens

AND

TO OUR FRIEND

Jos. A. Doyle

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PREFACE TO FOURTH EDITION

WITH the exception of a chapter on nitriding, added in the 1935 printing, this work has not been materially revised since 1927. Much new information has become available, and many of the old principles have received fresh verification in the last decade. Hence, for the volume to continue to serve its purpose, revision in the light of present knowledge became imperative. Since pressure of other work prevented the original author from taking an active part in the revision, he invited Battelle Memorial Institute to carry it out.

Battelle Memorial Institute considered that the previous editions of the book had served as an unusually effective link between theory and practice, largely because of the sound and practical point of view adopted in emphasizing the importance of the human element and in dealing with the various heating processes as a broad engineering problem. Believing that this point of view deserves perpetuation, and feeling that to aid in putting continued emphasis upon it is in line with the primary purpose for which the Institute was founded, i.e., helping to put science to practical everyday use in industry, Battelle accepted the invitation.

Among features in which major advances have been made since the last edition are, first and most important, an understanding of the significance of the grain coarsening tendencies of steels and developments in their control; the use of controlled atmospheres; the rise of high yield strength, low-alloy steels, insulating refractories for furnaces, forced circulation for more uniform heating in tempering, gas carburizing, and other processes; the use of gas-fired radiant tubes for large furnaces at moderate temperatures; the use of copper as an alloying element in steel; the normalizing instead of annealing of steel castings; and, in general, a clearer conception of the mechanism of heat treatment and of the specific effect of all the alloying elements.

These, together with the increase in special "tailor-made" steels, of too great importance to pass by entirely, necessitate the introduction of new chapters and, in order to conserve space, the condensation or elimination of other matter.

Even with the elimination of such matter as could be spared, the material that had to be added brought the bulk to too large an amount for a single volume. In the division into two volumes, those principles, processes, and fundamentals of control which relate to heat treatment of steel in general, whether it be plain carbon or alloy steel, were placed in Volume I, and the discussion of engineering carbon, simple alloy, complex alloy, and special-purpose steels, in Volume II.

Practically everything in these volumes has, naturally, been recorded in other books and in periodical literature. The excellent monographs in the "Alloys of Iron" series, the "Book of Stainless Steel," and other works of like completeness, including some put out by those who have a specific alloying metal to sell (which are valuable even if they do "put the best foot forward"), deal with each alloying element in a degree of detail possible in ample space. Valuable handbooks treat separate topics in tabloid form. The reader is especially urged to supplement this book by study of the appropriate sections of the "American Society for Metals Handbook." Symposia on special topics, held by technical societies, supply compact information in especially convenient form. Back of these compilations are thousands upon thousands of articles in which the information was originally presented. This very profusion, although of immense value when all available information upon a specific point is necessary, is a drawback when what is desired is a broad picture of the state of the art of the heat treatment of steels and an understanding of the fundamental principles involved, a picture which will enable the reader to apply the summation of known facts to the problems he personally faces.

Hence, the effort in this revision, as in the previous editions, is to present the whole topic from a consistent, unified point of view, in which the need of the practical man for pertinent information is never lost sight of.

ACKNOWLEDGMENTS

The acknowledgment to those who aided in the preparation of the previous edition, J. D. Cutter, J. A. Doyle, C. N. Dawe, J. V. Emmons, H. S. Foote, H. J. French, L. Garratt, A. W. F. Green, H. L. Green, M. A. Grossman, J. E. Harper, C. M. Johnson, A. H. Kingsbury, E. Larned, F. B. Lounsberry, A. F. MacFarland, J. A. Mathews, A. H. Miller, H. F. Moore, O. K. Parmiter, N. S. Stansel and T. W. Wickenden, made in the preface to the previous edition, is as heartily made here, since their cooperation helped build the framework of the present volume.

Added unpublished information or helpful comment on specific topics has been received from J. A. Doyle, A. J. Herzig, J. V. Emmons, C. H. Wills, and G. V. Luerssen, besides those to whom specific acknowledgment is made in the text.

Naturally, a primary source of information on advances in the heat treatment of steel is the publications of the American Society for Metals. The cooperation of the A.S.M. in permitting the use herein of any material appearing in any of its publications has materially facilitated the revision.

The Society of Automotive Engineers has similarly permitted the use of its data on S.A.E. steels, and other data have been taken with permission, from the publications of the American Institute of Mining and Metallurgical Engineers, of the American Society of Testing Materials, and from *Metals and Alloys*.

Permission for use of material appearing in "Nickel Alloy Steels," "Vanadium Steels and Irons," and "Bethlehem Alloy Steels," published, respectively, by the International Nickel Company, Vanadium Corporation of America, and the Bethlehem Steel Company, has likewise been an important aid.

The first draft of the revision, drawn up by H. W. Gillett, Battelle Memorial Institute, was read by the original author and by S. Epstein, Bethlehem Steel Company. The helpful comments of both have been utilized. The detailed study and constructive criticism of the whole draft by O. E. Harder and C. H. Lorig of Battelle deserve especial acknowledgment. Among the other members of the Battelle staff who have supplied information and criticism on particular chapters or topics are R. W. Dayton, H. C. Cross, B. W. Gonser, H. W. Russell, R. A. Sherman, and C. E. Sims. The index was prepared, references checked, and many tedious details attended to by Miss T. R. Reinberg. H. W. Gillett carried out the final selection and arrangement of the material.

All temperatures mentioned herein are in the Fahrenheit scale, and the F. is omitted, i.e., 1000° means 1000° F. For those who think in Centigrade, a conversion table is given on p. xiii. Mechanical properties are given in U. S. units, and, to save space, are abbreviated; e.g., instead of "Tensile strength, 100,000 lb./in.^{2}, Yield strength, 80,000 lb./in.^{2}, Elongation, 25 per cent in 2 inches, Reduction of area, 60 per cent, Izod impact 55 ft.-lb.", the notation "Tensile 100,000, Yield 80,000, Elong. 25%, R.A. 60%, Izod 55" will be employed. Similarly the name of an alloying element which recurs frequently in a discussion will not be spelled out; the chemical symbol will be used.

"Yield strength" is used throughout in place of "elastic limit" in accordance with the terminology of the American Society for Testing Materials.

The sections on nitriding are based upon the chapter written by V. O. Homerberg for the 3rd printing of the 3rd edition.

D. K. Bullens
Clyde E. Williams, Director
Battelle Memorial Institute

PREFACE TO FIRST EDITION

Modern Heat Treatment should be considered as an art or trade, since it certainly requires knowledge, skill and judgment for its proper performance. These, in turn, necessitate at least some knowledge of heat, of steel, and of the effect of heat upon steel. And all three factors are linked together by the "human element." The author has therefore endeavored to bring together the theoretical and practical sides of the general subject of steel and its heat treatment in such a manner as will, he hopes, be understandable by that "human element."

The theoretical or scientific side of heat treatment is not affected by a consideration of many elements which enter into the art or practical application of metallurgical principles, but unless performance in the shop compares favorably with the standard set in the laboratory, the value of the work is lessened. The technical principles have been dealt with heretofore more thoroughly than the application of these principles, and it is the purpose of this book more completely to consider this practical aspect in the hope that a better understanding of the factors governing commercial practice may suggest means of improvement in shop methods.

It has been the author's attempt to make the chapters dealing with the heating problem illustrate the necessity for considering all of the elements entering into the production of a uniformly heated product; of heat application rather than details of construction; of the importance of the human element and the relation it bears to the ultimate results; of encouraging a broad and common-sense view of the principles governing this important operation and which generally have been overlooked; and finally, of viewing the heating problem as an engineering proposition, adapting each fuel to proper furnace design and operation to meet the requirements of the problem in hand, and by so doing aim for the adoption of the standard heating unit in terms of finished product—"the cost of a unit of quantity of given quality."

He has attempted to make as practical as possible those chapters relating to steel and the effect of heat upon steel. Theories have been advanced only so far as has been thought necessary for a clear understanding of principles. Wherever possible, illustrations in the form of photomicrographs and charts have been given. The data given under the various types of heat-treated steels have been checked as far as possible and every effort has been made to be correct.

To the many friends who have aided him in the preparation of this book the author would express his sincere appreciation. Effort has been made to give due credit for cuts and data at the proper place, and for such as may not have been made, acknowledgment is hereby rendered.

Denison K. Bullens.

Philadelphia,
October 1, 1915.

TEMPERATURE CONVERSION TABLE

BY DR. LEONARD WALDO

Reprint from Metallurgical and Chemical Engineering

C. °	0	10	20	30	40	50	60	70	80	90
	F. °	F. °	F. °	F. °	F. °	F. °	F. °	F. °	F. °	F. °
−200	−328	−346	−364	−382	−400	−418	−436	−454		
−100	−148	−166	−184	−202	−220	−238	−256	−274	−292	−310
− 0	+ 32	+ 14	− 4	− 22	− 40	− 58	− 76	− 94	−112	−130
0	32	50	68	86	104	122	140	158	176	194
100	212	230	248	266	284	302	320	338	356	374
200	392	410	428	446	464	482	500	518	536	554
300	572	590	608	626	644	662	680	698	716	734
400	752	770	788	806	824	842	860	878	892	914
500	932	950	968	986	1004	1022	1040	1058	1076	1094
600	1112	1130	1148	1166	1184	1202	1220	1238	1256	1274
700	1292	1310	1328	1346	1364	1382	1400	1418	1436	1454
800	1472	1490	1508	1526	1544	1562	1580	1598	1616	1634
900	1652	1670	1688	1706	1724	1742	1760	1778	1796	1814
1000	1832	1850	1868	1886	1904	1922	1940	1958	1976	1994
1100	2012	2030	2048	2066	2084	2102	2120	2138	2156	2174
1200	2192	2210	2228	2246	2264	2282	2300	2318	2336	2354
1300	2372	2390	2408	2426	2444	2462	2480	2498	2516	2534
1400	2552	2570	2588	2606	2624	2642	2660	2678	2696	2714
1500	2732	2750	2768	2786	2804	2822	2840	2858	2876	2894
1600	2912	2930	2948	2966	2984	3002	3020	3038	3056	3074
1700	3092	3110	3128	3146	3164	3182	3200	3218	3236	3254
1800	3272	3290	3308	3326	3344	3362	3380	3398	3416	3234
1900	3452	3470	3488	3506	3524	3542	3560	3578	3596	3614
2000	3632	3650	3668	3686	3704	3722	3740	3758	3776	3794
2100	3812	3830	3848	3866	3884	3902	3920	3938	3956	3974
2200	3992	4010	4028	4046	4064	4082	4100	4118	4136	4154
2300	4172	4190	4208	4226	4244	4262	4280	4298	4316	4334
2400	4352	4370	4388	4406	4424	4442	4460	4478	4496	4514
2500	4532	4550	4568	4586	4604	4622	4640	4658	4676	4694
2600	4712	4730	4748	4766	4784	4802	4820	4838	4856	4874
2700	4892	4910	4928	4946	4964	4982	5000	5018	5036	5054
2800	5072	5090	5108	5126	5144	5162	5180	5198	5216	5234
2900	5252	5270	5288	5306	5324	5342	5360	5378	5396	5414
3000	5432	5450	5468	5486	5504	5522	5540	5558	5576	5594
3100	5612	5630	5648	5666	5684	5702	5720	5738	5756	5774
3200	5792	5810	5828	5846	5864	5882	5900	5918	5936	5954
3300	5972	5990	6008	6026	6044	6062	6080	6098	6116	6134
3400	6152	6170	6188	6206	6224	6242	6260	6278	6296	6314
3500	6332	6350	6368	6386	6404	6422	6440	6458	6476	6494
3600	6512	6530	6548	6566	6584	6602	6620	6638	6656	6674
3700	6692	6710	6728	6746	6764	6782	6800	6818	6836	6854
3800	6872	6890	6908	6926	6944	6962	6980	6998	7016	7034
3900	7052	7070	7088	7106	7124	7142	7160	7178	7196	7214
C. °	0	10	20	30	40	50	60	70	80	90

C. °	F. °
1	1.8
2	3.6
3	5.4
4	7.2
5	9.0
6	10.8
7	12.6
8	14.4
9	16.2
10	18.0

F. °	C. °
1	.56
2	1.11
3	1.67
4	2.22
5	2.78
6	3.33
7	3.89
8	4.44
9	5.00
10	5.56
11	6.11
12	6.67
13	7.22
14	7.78
15	8.33
16	8.89
17	9.44
18	10.00

EXAMPLES: 1347° C. 2444° F. + 12°.6 F. = 2456°.6 F.: 3367° F. = 1850° C. + 2°.78 C= 1852°.78 C.

CONTENTS

VOLUME I. PRINCIPLES, PROCESSES, CONTROL

SECTION I. METALLURGICAL PRINCIPLES

CHAPTER PAGE

1. Definitions—Testing of Steel 1
2. The Structure of Steel 74
3. General Annealing Practice 94
4. Spheroidizing—Tool-Steel Annealing 124
5. Normalizing and "Austempering" 131
6. Principles of Hardening 147
7. Tempering and Toughening 186
8. Precipitation Hardening 203
9. Grain Size 219

SECTION II. SURFACE-REACTION PROCESSES

10. Case Carburizing 251
11. Heat Treatment of Carburized Steel 280
12. Controlled Atmospheres 292
13. Cyaniding, Nitriding, Surface Hardening, Coating 342

SECTION III. CONTROL OF HEAT-TREATING OPERATIONS

14. Proper Heating in Forging and Rolling Practice 375
15. Sources of Effective Heat for Heat-Treating Operations—Fuels and Furnaces 391
16. Application of Heat in Heat Treating 412
17. Control of the Cooling Phase 427
18. The Human Element 429

Index 435

STEEL AND ITS HEAT TREATMENT

SECTION I. METALLURGICAL PRINCIPLES

CHAPTER 1

DEFINITIONS—TESTING OF STEEL

Introduction. Steel combines strength, workability, and cheapness to a degree unparalleled in any other material of construction. It is the mainstay of engineering, indispensable alike in peace and war.

The engineer would be badly cramped in the utilization of steel were his use of it confined to those purposes sufficiently well served by steel as it comes from the rolls, the forge, or the mold. Improvement in, and gradation of, the properties of steel, conferred by controlled heating and cooling, fit heat-treated steel for myriads of uses which raw steel will not satisfactorily serve.

Finer and finer gradations of properties are being, and will continue to be, demanded by the user of steel. Still closer control of steel-making processes and of heat-treating methods must be developed. To this end accumulated practical experience, improved equipment, and increased understanding and application of the fundamental principles involved must all be utilized. Men, machinery, and knowledge make a stool which needs all three of its legs to stand on.

Vast strides have been made in recent years in the way of equipment. Processes that were previously workable only in the laboratory by application of the greatest skill have become the commonplaces of production through the development of special equipment and of control devices, far more than human in their action. Continuous and automatic processes are in vogue that perform operations with a speed and certainty that far surpass the skill of the artisan, and maintain a uniformity of quality unthought of when only manual dexterity was involved. But the planning of such processes involves exactly the same metallurgical principles as were utilized in non-mechanized processes. The human element is still involved. Precise knowledge is even more emphatically demanded than it was when the properties of the product were not so rigorously specified.

Definitions: Steel, Alloy Steel. Steel is defined as an alloy of iron and carbon, usefully malleable as cast. The phrase "malleable as cast" serves to exclude gray cast iron, white cast iron, and malleableized cast iron. For practical purposes one might take the lower carbon limit of the low-carbon steel, No. 1010 in the Society of Automobile Engineers' series, i.e., 0.05%, as the dividing line between iron and steel.

The practical upper limit is not so clearly defined. Save for tool steels, there are few steels that contain more than 1% carbon, and in these 1.25% is about the upper limit. The new "graphitic" silicon steels may have as much as 1.50%. From the point of view of theory, an upper carbon limit of 1.7% corresponding to that point in the iron-carbon equilibrium diagram beyond which iron carbide cannot be wholly held in solid solution at any temperature, is often linked with the definition of steel. Although common terminology disregards this limit in some alloy steels that have a different type of carbide, e.g., high-carbon, high-chromium steels, the limit does serve as a first approximation.

The lower limit of the carbon content might be fixed as that value below which iron carbide cannot be thrown out of solid solution at any temperature. This is lower than 0.01% so practically all commercial irons are to some slight degree amenable to heat treatment and could be considered as steels. Common parlance tends to call the very low-carbon alloys, that are only slightly or with difficulty affected by heat treatment, "irons" rather than "steels."

Actually, a truly pure alloy of iron and carbon alone would behave quite differently from a commercial carbon steel of equal carbon content because of the intentional addition to true steel of manganese, silicon, or some other "deoxidizing" element, added to counteract the tendency for iron oxide to be present along with iron and iron carbide. These added elements are considered as normal constituents of carbon steel until their amounts are raised to such proportions that deoxidation is no longer the main purpose and some specific effect is obtained by the addition, when an "alloy steel" results. Traces, at least, of sulphur and phosphorus and generally of other elements, not intentionally added but arising from the iron ore and the scrap used as raw materials, will also be present.

Elements Occurring in Steel. At least a score of the chemical elements are usefully employed in steels of different kinds. In metallurgical, as well as chemical literature, these elements have to be mentioned constantly. Their names are abbreviated to the chemical symbols. The heat treater always needs to know the chemical composition of the steel to be treated. Since the reports of the chemical analyses are universally given without spelling out the full name of the element, the abbreviations must be learned.

These abbreviations will be used in the text in order to save space. Following are the names of, and symbols for, the steel-making elements:

Iron	Fe	(Latin, *ferrum*). The base of all steels.
Carbon	C	A non-metal.—The element which most influences steels in respect to heat treatment.
Silicon	Si	A small amount is present in all ordinary killed steels.
Manganese	Mn	A small amount is essential in all ordinary steels.
Phosphorus	P	A non-metal, but alloys with iron.
Sulphur	S	A non-metal, present as sulphides.
Oxygen	O	A gas, present as oxides.
Nitrogen	N	A gas, present as nitrides.
Hydrogen	H	A gas, present, probably as the element, in most steels, but only in a very tiny amount.

The elements listed above are present in so-called "plain carbon" steels. The amount of Si, Mn, P, or S may be increased beyond the normal in order to confer special properties, when they class as alloying elements.

The other important metallic alloying elements are:

Nickel	Ni	
Copper	Cu	(Latin, *cuprum*)
Chromium	Cr	
Molybdenum	Mo	
Tungsten	W	(Latin, *wolfram*)
Vanadium	V	

Other metals used in special alloy steels or added for particular purposes are:

Aluminum	Al
Cobalt	Co
Titanium	Ti
Columbium	Cb
Zirconium	Zr

Non-metals occasionally used for purposes similar to that when S is utilized as an alloying element are:

Selenium	Se
Tellurium	Te

A few more elements may accidentally occur in, or be intentionally added to, steel, but in such small amount or so rarely as to need no comment here.

It will be noted that the chemical symbols follow naturally from the English names except in the cases of iron itself (Fe), Copper (Cu), and Tungsten (W).

Heat Treatment. "Steel" is a very large subject. In this book, steel will be considered only in relation to its heat treatment. The processes of steel-making will not be dealt with, nor shall we be concerned with strengthening by cold-working processes. Steels that are not amenable to heat treatment will be passed by with scant attention, since they fall outside the scope of our subject.

"Heat treatment of steel," in its broadest sense, refers to any process involving heating and cooling of the solid metal by which the properties of the steel are altered without altering its average chemical composition. In a more limited sense, it applies especially to the quenching and tempering (hardening and toughening) processes, so that the term "heat-treated steel," without further qualification, is applied to steel that has been cooled with a certain degree of rapidity to produce changes in properties and then reheated to produce still other changes. However, the single operations of cooling at a specified rate, as in "normalizing," or of heating, as in the annealing of a cold-worked steel, come under the broad definition.

The object of heat treatment is to produce mechanical or physical properties that make the steel better adapted for industrial use.

Objects of Testing. In order that producer and consumer may speak a common language in discussing the quality of steel, it is necessary that "quality" be described, not merely in vague or qualitative phrases but, wherever possible, in definite figures. This leads to mechanical and physical testing, first for the selection of suitable material, second for its duplication. The effects of heat treatment upon steel of given chemical composition as well as the proper correlation of composition and treatment are appraised by such testing.

It will be necessary throughout this book to refer constantly to quantitative properties, as determined by testing, in order to describe the effects produced by choice of steels and modifications in heat treatment. Some discussion of testing, therefore, is in order.

The heat treater is interested in tests that will enable him to tell the steel maker what he wants in the steels he uses, and in those tests which enable him to satisfy the user of the heat-treated part that it has the properties which the user, in his turn, demands. In both cases definite specifications are more helpful than vague phrases, provided always, that the specifications really define the properties actually desired.

Tests upon Material for Heat Treatment. Chemical analysis is the primary specification in the selection of steels for heat treating.

Not only the properties of the finished part but also the correct temperatures for treatment depend on the composition. If a given temperature cycle is to be employed, the composition cannot vary too widely, else uniformity of properties will be unattainable. The steels most used for heat treating are designated by the Society of Automotive Engineers by numbers, which, in a convenient code (see Chap. 10, Vol. II) convey information as to the median composition desired, but for each element so specified there is a certain range of allowable composition. For C, this range is usually 0.10%, i.e., ten "points." Such a variation is permitted because of the difficulty of producing successive heats of steel of exactly the same content in carbon and other elements. If too close a range were specified, the steel maker would have to select heats and divert the off-heats to other uses; and this would increase the cost. Closer specifications are frequently imposed when the purchaser believes that the extra cost is repaid by the greater uniformity of properties in the finished part. The heat treater can usually adjust his temperatures to compensate approximately for reasonable variation in composition, but this involves keeping each lot of steel separate, and correctly identified, as well as altering the heat-treating schedule to correspond.

It must be remembered that errors may creep into sampling and into the chemical analysis itself, so that a small leeway has to be allowed the analyst. An analysis for the elements usually determined in a C steel, that is, C, Si, Mn, S, P, may be inadequate, since residual metals, especially Ni, Cu, and Cr, may also be present in sufficient amounts to alter the properties of the steel. Although there usually is not enough of these elements to affect heat treatment or properties, the absence of larger amounts cannot always be assumed.

Constituents not included in routine chemical analysis but known or suspected to be of major importance are O, N, and H. Nitrogen can be determined easily and accurately; but O, which is present as oxides, silicates, etc., in steels suitable for heat treating, in amounts varying from, say, 0.02 to 0.002%, is more difficult to determine, as is shown by the summary of Thompson, Vacher, and Bright.* This is especially true, since it is not the total O but the relative amount of the different oxides, silicates, etc., in which it is combined that counts the most. Hydrogen, although only around 0.001% is usually present, is nevertheless held responsible for some of the troubles of "flaky" steels met in heat treatment.[129]

The "spark test," although of no value for exact analysis, is useful in separating an accidentally mixed stock of different alloy steels. Detailed discussions of this are readily available.[2]

---

* References are at the end of each chapter.

Besides the variation in average composition that occurs from heat to heat, there is likely to be segregation in any given ingot, so that not every point in the bars or other shapes made from it has exactly the same composition. The worst of the segregation is eliminated by cropping the ingot, and adequate cropping is essential to a product of high chemical uniformity. Segregation, of course, is the less the smaller the ingot, and in tool steels, for example, the production from small ingots is preferred to that from large ones.

Examination under the Microscope. Segregation of S can be revealed by sulphur printing. Segregation of P, resulting in a banded structure in the steel, is revealed by polishing, etching, and microscopic examination. Microscopic examination will also show whether the steel is in the desired structural condition.

The presence of excessive amounts of non-metallic inclusions is shown by microscopic examination of unetched polished sections.

Inclusions. There are a variety of "inclusion charts" by which users make a "cleanliness rating" of steels for severe duty. One such series of charts and method of rating is described by Epstein.[3] Some used abroad have been described by Diergarten[4] and Perrin.[5, 6] Each steel company has to work to different sets of cleanliness standards set up by different customers and usually has its own internal standards as well. Committees of the American Society for Metals and of the Society of Automotive Engineers are studying the problem of unification of standards. The problem is a difficult one because of the fact that inclusions are not necessarily distributed uniformly and the polishing and examination of many—even hundreds—of specimens may be necessary to sample a given lot of steel adequately.

A different method of attack is suggested by Hare and Soler,[6a] who modify methods that had been previously discussed in the literature. They advocate separation of the inclusions by controlled electrochemical solution of the steel matrix and determining the weight percentage. The residue is itself examined with a petrographic microscope and by X-ray diffraction methods and then analyzed chemically in order to find in what mineralogical form the non-metallic inclusions exist. McCombs and Schrero[7] assembled a bibliography of more than 2000 references dealing with inclusions.

Whatever method of evaluation is used, there remains the primary problem of what limits of cleanness should be set. This requires much engineering judgment in order to avoid failures in service from the effect of inclusions, on the one hand and, on the other hand, to keep from imposing standards so difficult to meet that the cost of the steel rises to a figure that cannot be paid. The standard imposed by the makers of an aircraft engine would be more severe than that ordinarily needed by others. In this connection it is well to recall that many of the properties of free-machining steels very high in sulphide inclusions are hardly affected to a determinable degree by the presence of the inclusions, and that wrought iron is not barred from industrial service by its large slag content. In general, soft, low-C steels are rather unaffected by inclusions whereas hard, high-C steels, or other steels heat treated to tool hardness or spring temper, may be detrimentally affected. A valuable discussion of the general inclusion problem is given by Larsen.[8]

Deep Etching. Ingot features that affect quality are the presence of concealed pipe or spongy centers in improperly fed ingots, a fault preferably avoided by the use of big-end-up, hot-topped ingots. Unless the ingot has been cast at the proper temperature, properly heated for rolling or forging, and has had a sufficient degree of hot working, traces of the coarse, dendritic, cast structure may persist. Deep etching is helpful in detecting such defects.

Deep etching consists in attacking a specimen, usually a sawed-off cross section, by acid, usually about half commercial hydrochloric acid (perhaps with a small amount of sulphuric acid) and half water, at about 160° for 15 min. to 1 hr., depending upon the type of steel.

Seams, pipes, spongy centers, large slag inclusions, and the like are enlarged and shown up in accentuated fashion by the acid attack. If the dendritic structure of the original ingot has not been broken down in the working of the steel, that structure may be revealed, though whether such a structure is undesirable is a question only to be answered in respect to a specific steel and a specific service. Gross imperfections that exist but cannot be seen without etching are shown up, and when these would be detrimental in service the steel may be discarded and the work may be avoided that would otherwise have been put on it, only to have it fail to pass final inspection (or still worse, to slip through such inspection and come back, failed in service). The test so exaggerates and magnifies inhomogeneities that the unskilled may be tempted to reject normal, useful material on the basis of its looks after deep etching; so the test must be employed with discretion tempered by experience. Suggested practice for the test[9] and extended discussion of its application,[10] are available in the literature.

Critical Temperatures. The temperature to which the steel must be heated in order to harden on quenching can be estimated approximately from the chemical composition, be determined by a "critical point" test, or be found by attempting to harden it from a series of temperatures around that calculated from its composition (see Chap. 2).

Hardenability and Grain Size. The ease of hardening of a steel depends primarily upon its composition, but also to some degree upon its structure, which depends upon its prior treatment; and, unless the incoming steel already has proper structure, a normalizing treatment, i.e., air cooling from above the critical temperature, may be required to produce it.

Besides this, steel may vary in hardenability from other causes. The depth of hardening may be very different in steels of closely similar composition and prior treatment. This variation is related to the grain size produced at elevated temperature, and various grain size and hardenability tests are in vogue for detecting such variations in quality before selecting the exact heat treatment for a given lot of steel. These will be discussed in Chap. 9.

Tests upon Heat-treated Material. The tests for evaluation of the success with which the heat-treatment process has been carried out, and by which the heat treater may both control the process and prove that he has met the specifications, as well as those by which the engineer tries to appraise the suitability for service, are in general quite thoroughly standardized and readily expressed in quantitative figures. Two general classes of tests are used: static tests in which stress is applied slowly and only once, as in tensile or bend testing; and dynamic tests, in which the load is applied rapidly, as impact testing, or repeatedly, as in endurance testing. Of the static tests, tests for hardness are the simplest, cheapest, and most widely used for process control.

Hardness Testing. Hardness is resistance to penetration. It is most often commercially measured by pressing a harder object into the metal to be tested and noting the penetration. For soft steels the penetrators are usually hardened steel balls; for harder ones, tungsten carbide balls,[11] or more commonly, diamond points, either cones or pyramids. Complete discussions of hardness testing are given in several monographs.[12, 13, 14]

Brinell Hardness. The Brinell Hardness Number (abbreviated to B.H.N., or simply Brinell) is calculated from the diameter of the impression left after a 10-mm. diameter ball (supposed to be rigid and not itself to deform) is pressed into the specimen vertically in not less than 6 sec. and then held at the maximum load of 3000 kg. for 30 sec. The Brinell number is calculated as the load divided by the spherical area of the impression. The harder the steel, the smaller the indentation for a given load, and the higher the Brinell. For many purposes the indentation diameter can be used, without transferring this into the Brinell number. Table 1 (p. 10) gives Brinell numbers calculated from the diameters. The diameters, the load, and the time are all capable of accurate measurement, i.e., the method is not dependent on anything that cannot be quite readily checked.

For steels of 70 to 500 Brinell, hardened, accurately ground and polished 10-mm. balls of steel of the regular ball-bearing type are commonly used. Use on harder steels permanently deforms the ball itself; the first impression made with the ball gives a hardness lower than the true one and in subsequent use the results become even more inaccurate. Work-hardened balls, which decrease this error, are obtainable, with which little inaccuracy results up to 600 Brinell, and carbide balls can be had that will serve to around 800.

For very soft materials, a 500-kg. load is used. Brinell numbers can be computed by the formula $BHN = \dfrac{P}{\frac{\pi D}{2}\left(D - \sqrt{D^2 - d^2}\right)}$ where P is pressure in kilograms, D and d, respectively, the diameters of the ball and of the impression in millimeters.

So-called "Baby Brinell" tests with tiny balls are sometimes made upon objects too small for the standard test. The data computed from non-standard conditions usually differ somewhat from those obtained under standard conditions, and, when any load, ball, diameter, or ball material other than the usual ones is employed, the fact should be stated.

Precautions [15] necessary in Brinell testing are to see that the loading device shall press the ball upon the specimen in a truly vertical direction, to observe the required rate and time of loading, and to make sure that the maximum load is reached but not exceeded. That is, the Brinell machine must be kept in calibration within an error of ± 10 kg. in 3000. The ball must have the nominal diameter (10 mm.) within 0.025 mm. (0.001 in.) and be discarded if deformed beyond that limit. The measuring microscope must read true dimensions to 0.01 mm.

The diameter of the impression must be determinable within an error of 0.01 mm. This requires that the specimen be flat, smooth, and freed from deep tool marks; therefore smooth machining, filing, or grinding is often necessary. If the specimen is curved, the radius of curvature should be at least 25 mm., (1 in.) and the diameters of the impression must be averaged, since it is not quite circular. The center of the impression must be at least three of its own diameters away from the edge of the specimen or the center of another impression.

Brinell Hardness Numbers

Diameter of Steel Ball = 10 mm.

Diameter of Ball Impression, Mm.	Hardness Number for a Load of 3000 Kg.	Diameter of Ball Impression, Mm.	Hardness Number for a Load of 3000 Kg.	Diameter of Ball Impression, Mm.	Hardness Number for a Load of 3000 Kg.	Diameter of Ball Impression, Mm.	Hardness Number for a Load of 3000 Kg.	Diameter of Ball Impression, Mm.	Hardness Number for a Load of 3000 Kg.
2	946	3	418	4	228	5	143	6	95
2.05	898	3.05	402	4.05	223	5.05	140	6.05	94
2.10	857	3.10	387	4.10	217	5.10	137	6.10	92
2.15	817	3.15	375	4.15	212	5.15	134	6.15	90
2.20	782	3.20	364	4.20	207	5.20	131	6.20	89
2.25	744	3.25	351	4.25	202	5.25	128	6.25	87
2.30	713	3.30	340	4.30	196	5.30	126	6.30	86
2.35	683	3.35	332	4.35	192	5.35	124	6.35	84
2.40	652	3.40	321	4.40	187	5.40	121	6.40	82
2.45	627	3.45	311	4.45	183	5.45	118	6.45	81
2.50	600	3.50	302	4.50	179	5.50	116	6.50	80
2.55	578	3.55	293	4.55	174	5.55	114	6.55	79
2.60	555	3.60	286	4.60	170	5.60	112	6.60	77
2.65	532	3.65	277	4.65	166	5.65	109	6.65	76
2.70	512	3.70	269	4.70	163	5.70	107	6.70	74
2.75	495	3.75	262	4.75	159	5.75	105	6.75	73
2.80	477	3.80	255	4.80	156	5.80	103	6.80	71
2.85	460	3.85	248	4.85	153	5.85	101	6.85	70
2.90	444	3.90	241	4.90	149	5.90	99	6.90	69
2.95	430	3.95	235	4.95	146	5.95	97	6.95	68

The specimen, for test under the 10-mm. ball and 3000-kg. load, must be at least 0.4 in. thick. These precautions are more fully described in several accessible sources.[16, 23]

"Cloudburst" Hardness Testing. To reveal soft spots on a hard surface, a shower of steel balls may be dropped on the specimen or, in production control, on an actual part. By regulating the height of fall, the balls can be made to mark the soft spots without appreciably indenting the hard ones. Such a test may be looked upon as a special case of shot blasting. Details are discussed in the "National Metals Handbook."[16]

Actual shot blasting is advocated by Peterson[17] for inspection of high-speed steel tools and coil-spring material.

Mutual Indentation Hardness Tests. Hardness measurements can be made by pressing two suitable specimens of the same material together, and calculated into Brinell values instead of using a hard ball or other penetrator. Crossed prisms or cylinders are sometimes used;

but two cylinders, lying parallel and in contact throughout their length, one above the other, held in suitable blocks to maintain alignment, are more commonly used. The pressure applied flattens the portions of the two cylinders in contact. Measurement of the flat areas together with the pressure applied gives data from which Brinell hardness may be computed. Cowdrey [18] has described the method in detail.

Hot Hardness. The mutual indentation method is seldom applied in ordinary hardness testing. It finds chief application to determinations at elevated temperatures. Harder and Grove [19, 20] developed and described the method for high temperatures. It has been successfully used in the study of high-speed steels, valve and die steels, and other materials for high-temperature service.

An alternative method is to use the regular Brinell test with a ball that does not deform appreciably at the test temperature. Tungsten carbide balls have been so used by several workers.

Vickers Pyramid Test. An indentation hardness test applicable over a wider range of hardnesses than the ball hardness method is the diamond pyramid test, used in the Vickers and some other instruments, results of which are most commonly reported as Vickers or Vickers-Brinell numbers. The point of a small diamond pyramid (136°) with a square base is impressed into the specimen just as is done in Brinell testing, but under a very much lighter load.[21] The impression is much smaller than in the Brinell test; hence, much thinner specimens may be tested. The impression tends to be more sharply outlined and its diagonal more easily measured than a spherical one, though on account of its small size a higher magnification is required. The hardness number is calculated, as in the Brinell test, from the load and area. The diamond point does not deform plastically and only slightly elastically, so that the calculated hardness is not appreciably affected by use of different loads; hence the depth of indentation can be regulated to suit the thickness of the specimen. It need be only of the order of 0.002 in. The means of applying the load can, of course, be any suitable device, whether the Vickers machine itself or not.

The formula for calculation of Vickers hardness is:

$$\text{Hardness} = 1.8544 \frac{\text{load}}{(\text{diag.})^2},$$

the load being measured in kilograms and the diagonal of the impression in millimeters.

The advantages of the Brinell and Vickers-Brinell methods, in that all factors can be measured in absolute units, are somewhat lessened by the necessity of measuring the impression, which slows up the operation.

Devices are now on the market by which the image of the Brinell or Vickers impression is projected on a ground-glass screen, against a scale, for quick reading.

An advantage of the pyramid method is that, by varying the loads, hardness determinations may be made on the softest and the hardest alloys, and the results are expressed on a single scale without the necessity for shifting from one scale to another as in most of the other methods.

Rockwell Cone Test. To hasten the testing, a method somewhat similar to the Vickers has been devised in which a diamond cone (120° with a slightly rounded tip), instead of a pyramid, is used as penetrator; but the depth of penetration is measured by the testing instrument, instead of waiting to measure the breadth of the impression. This is the Rockwell cone hardness tester. In the Rockwell test the penetrator is first seated by a small (so-called "minor") load, the dial set to zero, the larger (major) load (regularly 150 kg.) applied for a specified time, the major load removed, and the hardness number is then read directly from the dial.

Each instrument has to be separately calibrated in arbitrary units, and any difference in the shape of the penetrator tip will affect the calibration. This type cannot be so certainly checked in absolute units as can the Brinell or diamond pyramid types but has the advantage of great rapidity, a reading being made in a few seconds.

The depth of penetration is of the order of 0.005 in., so that thin specimens may be tested. In order to make it feasible to test still thinner hard surfaces, like those on nitrided steel, a similar piece of apparatus, the Rockwell superficial hardness tester, has been devised in which the depth of penetration is only about 0.002 in. and the depth-measuring device is correspondingly more sensitive. Figure 1 shows the relative depths of penetration.

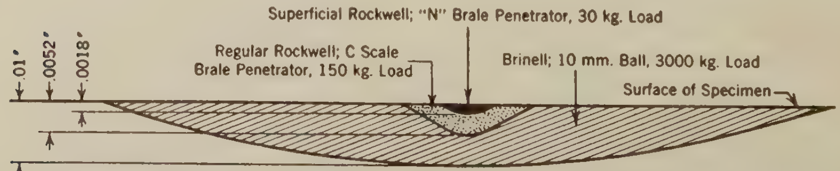
Superficial Rockwell; "N" Brale Penetrator, 30 kg. Load
.01"
.0052"
.0018"
Regular Rockwell; C Scale Brale Penetrator, 150 kg. Load
Brinell; 10 mm. Ball, 3000 kg. Load
Surface of Specimen

Fig. 1.—Relative depths of Brinell, Rockwell cone, and Rockwell superficial hardness impressions.

The Rockwell testers can be used with different minor and major loads, and these have been varied for work on material of different hardness ranges and scales, differently lettered, provided for such cases. Hence, the scale used must always be stated. Unless a different scale is stated, the "C" scale, the regular penetrator, and the regular major load of 150 kg. are implied.

Besides the cone penetrator, the Rockwell machine is provided also with a $\frac{1}{16}$-in. diameter steel ball, used with a major load of 100 kg. This gives a type of "Baby Brinell" test more rapidly carried out, because measurement of depth is done by the machine itself. The Rockwell "B" test is only employed on soft steels, not harder than 300 Brinell. Detailed descriptions of the Rockwell method are available.[22, 23]

Shore Rebound Test. Another type of diamond-pointed indenter and a different method of measurement are used in the Shore Scleroscope. A diamond-tipped hammer is dropped from a definite height upon the surface to be tested, and the hardness is measured by the height of the rebound, the scale being graduated in arbitrary units. The depth of penetration is very small, so that the instrument can be used on case-hardened specimens. The rebound is regulated more by the elastic limit of the specimen than by its tensile strength, so that the Shore does not measure exactly the same type of hardness as do the indentation methods. The scleroscope requires a very smooth surface on the specimen, which must be solidly held, in a horizontal position, and the hammer must fall exactly vertical. Different instruments tend to give slightly different readings, and it is important to check the readings often upon standard blocks. Several Shore readings should be taken and averaged, care being taken not to let the hammer fall twice in the same place.

The Shore instrument was once widely used upon steels too hard for good Brinell readings, but with the advent of the diamond cone and pyramid penetrators its use has been less common. It is especially useful for testing large chilled rolls, too large to take conveniently to the other types of equipment, for the Shore instrument can be transported to the work.

General Precautions in Hardness Testing. Certain precautions are necessary with any of these hardness tests. The load must be applied truly perpendicular to the specimen to be tested. The surface should be flat and all scale or decarburized surface removed, the latter precaution being the more important the shallower the indentation. The surface resting on the anvil should make good contact so as to avoid springiness in thin specimens. Curved surfaces of fairly large radius of curvature in comparison with the width of the indentation may be tested, but results will not be exactly the same as on a flat surface. With the Brinell test, the long and short diameters of the impression on a curved surface are measured and averaged. The smaller the indenter, the more carefully smoothed and polished must the specimen be, the 10-mm. Brinell ball being the least rigorous in its requirements, the scleroscope the most.

No diamond point can be expected to give true results if it is chipped, and precautions as to vertical loading, smoothness of applying the load, etc., are needed for the sake of the diamonds as well as for accuracy. All hardness testers should be periodically checked against standard blocks of known hardness and uniformity and the instruments kept in calibration. Those that have empirical scales, such as the Rockwell and scleroscope, should be returned to the makers when out of adjustment, since they are not readily calibrated by the user.

In all hardness testing the specimen must be firmly supported. This, of course, is especially vital for thin specimens. Very great attention has to be given to the "anvil effect" in testing very thin specimens, and there is much discussion in the literature as to the limiting thickness for accuracy in the various types of test, especially upon soft sheets. In heat treatment the objects are usually massive enough so that no trouble is encountered, though shallow carburized or nitrided cases involve analogous problems.

In careful work with any of these types of hardness testers, several readings, usually four, are made and averaged.

File Hardness. Besides the quantitative hardness tests, a much-used and useful shop test is filing with standard testing files. The personal equation of the man handling the file, the angle at which it is held, the force applied, etc., are variables; but, by frequently checking himself by filing standard blocks and discarding test files as soon as they become worn, the tester can make accurate comparisons of the hardness of steels, from say 60 Rockwell "C" up. The usual testing files will not "touch" quenched steel of 64-68 Rockwell "C" but will bite quenched steel that has been tempered at 375° to 64-66 Rockwell "C." It is possible to make files that will cut hardened steel tempered at 400° and not that tempered at 375°, and still further gradations are possible.[24, 25]

Scratch Tests. A type of file testing may be carried out by scratching with a sharp needle point, the point usually being ground on the tip of a square, tapered file made from plain C steel with a hardness, as quenched, of 65 Rockwell C. By tempering at a series of temperatures, needles of a graded range of hardness can be obtained. This method is described by Hamilton[26] and has been proposed[27] for consideration as an A.S.T.M. standard method, for use on materials or surfaces so thin that other methods are not applicable. More elaborate scratch tests are made by dragging a diamond or sapphire point under a definite load, over a surface, polished as for metallographic examination, then measuring the width of the scratch. A device for this test is sold under the name "Microcharacter." The method is used in research work to differentiate between microconstituents of an alloy. It is suggested by Conley and co-workers[28] as of possible applicability to the study of carburized or nitrided cases and, since the test is itself one of cutting, to evaluation of machinability. The method is seldom applied for commercial evaluation of hardness.

Pendulum Hardness. Another type of hardness test equipment, devised by Herbert, consists of an apparatus with a hard ball near the center of gravity of a heavy framework. The whole apparatus is rested on the specimen through the ball, thus indenting the specimen. When the apparatus is tilted and released, it swings like a pendulum. The angle of swing after release and the time for a definite number of oscillations are alternatively taken as measures of hardness. Claims have been made that these values give information not brought out by other testing methods. Much care and skill are required to get concordant check readings. This device has found practically no commercial use in this country.

Many other types of hardness testers have been devised for routine or research work, some of which are as useful as those mentioned above, others useful only in special cases.

Hardness Conversions. It would be highly desirable to be able accurately to convert hardness figures obtained by one method into those obtained by another. Many investigators have made such comparisons on the machines available to them and have given conversion formulae, tables, or curves. Any one investigator is likely to get quite smooth curves, applicable to his own machines, as long as they are in calibration and as long as the types of steels compared do not vary too widely. The results of different investigators are prone to deviate considerably, especially in the higher hardness range. When such different results are plotted, the curve becomes a band rather than a line, the width of the band representing the degree of uncertainty likely to be encountered among commercial machines in good repair. Nevertheless, the average curve (heavy lines in Figs. 2-6) obtained from all these data affords a convenient conversion, usable for general purposes. In preparing Figs. 2-6, data from many sources were plotted. The bands show the range of reasonable values recorded, with the omission of a few data that were obviously obtained from instruments in poor repair. For conversion tables to Monotron, Durometer, Herbert Pendulum, and Duroscope, see Hruska.[29] A single chart for conversion, one to the other, of average curves for Brinell,

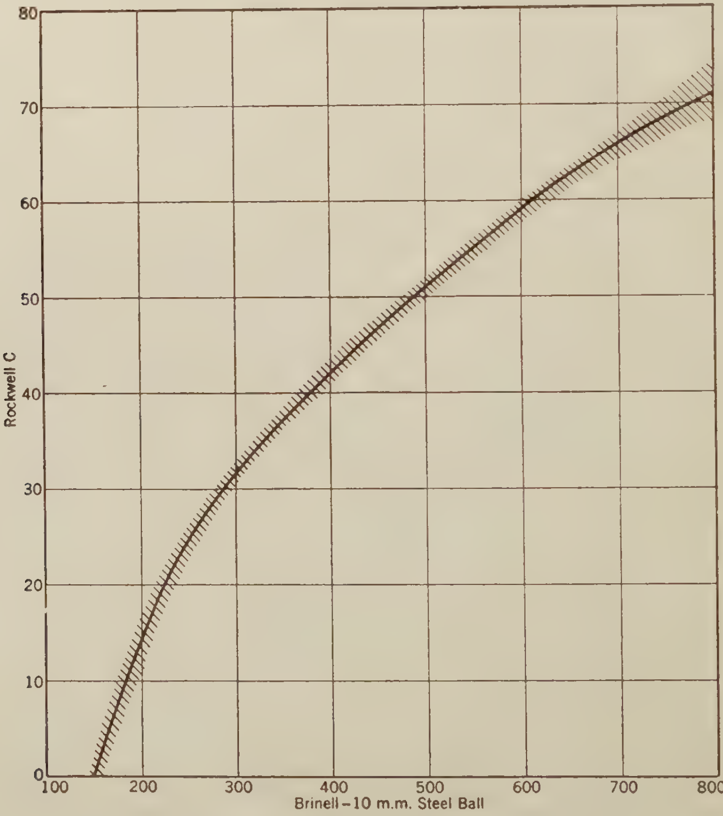

Fig. 2.—Conversion chart. (Brinell and Rockwell C.)

Vickers, Monotron, Shore, and Rockwell (C, B, and E scales) has been published.[30] Many conversion charts are shown in a recent German publication,[124] which discusses hardness testing in great detail, with references to the world literature.

The fanning out of the Brinell-Rockwell conversion curve of Fig. 2 above 600 Brinell indicates the difficulty of making good conversions at high hardness.

The Vickers to Brinell curve, Fig. 3, shows that the numbers in the two systems are the same up to 300. Above this the Vickers method gives the higher number. Figure 3 also shows the difference between the use of a deformable steel ball and a rigid ball in the Brinell test, and explains why the steel ball is unreliable above 500 Brinell.

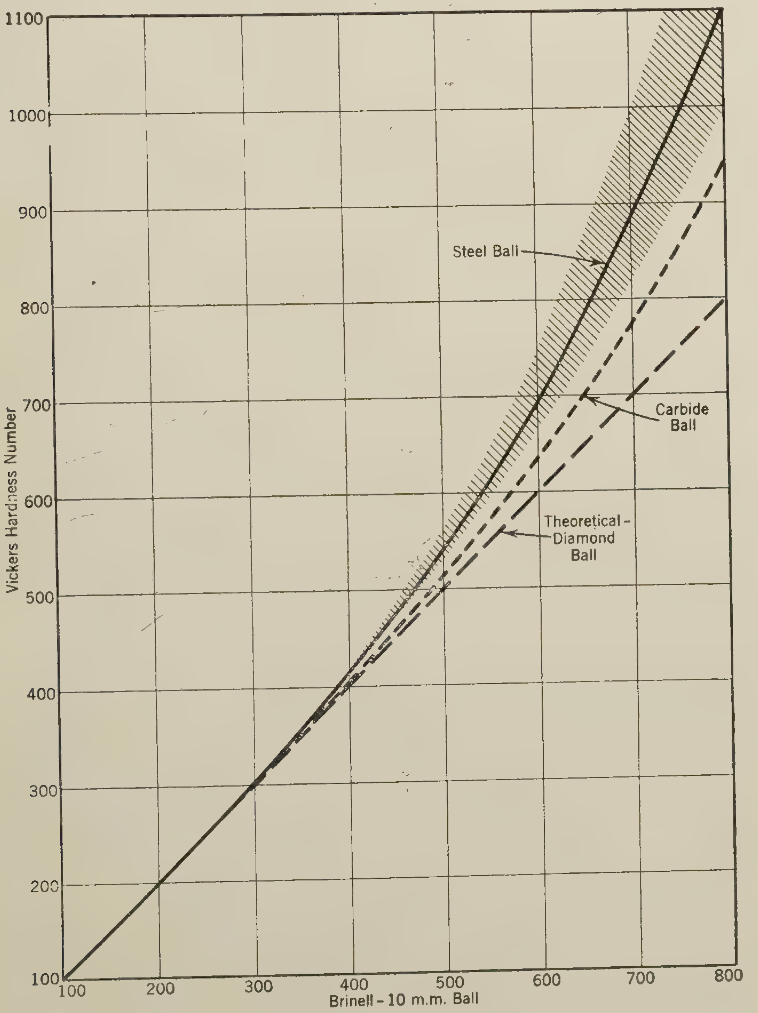

Fig. 3.—Conversion chart. (Brinell and Vickers)

Figure 4, Rockwell to Vickers, shows fair convertibility up to 50 Rockwell C but a very poor degree of certainty at high hardness, since, according to different investigators, 68 Rockwell C may mean anything from 875 to 1200 Vickers.

Figure 5, Shore to Rockwell, shows increasingly poor correlation below 60 Shore, and even above that the conversion is in doubt by some three to five points on either scale.

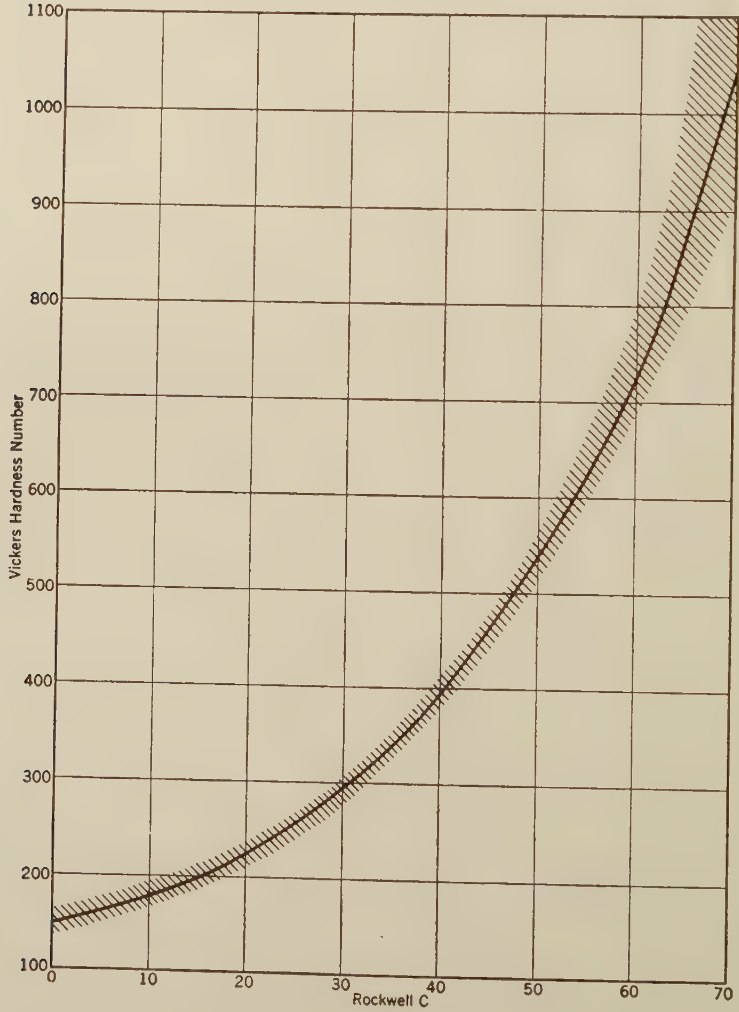

Fig. 4.—Conversion chart. (Rockwell C and Vickers)

The shaded band shows the deviations ordinarily met. Hengemühle and Clauss [31] cite even greater deviations among 7 "original" Shore instruments in use in German laboratories and studied by them.

The range of readings on these seven instruments is shown in Fig. 5 as the area *ABCD*. These authors discuss in detail the various factors that may affect a rebound hardness determination.

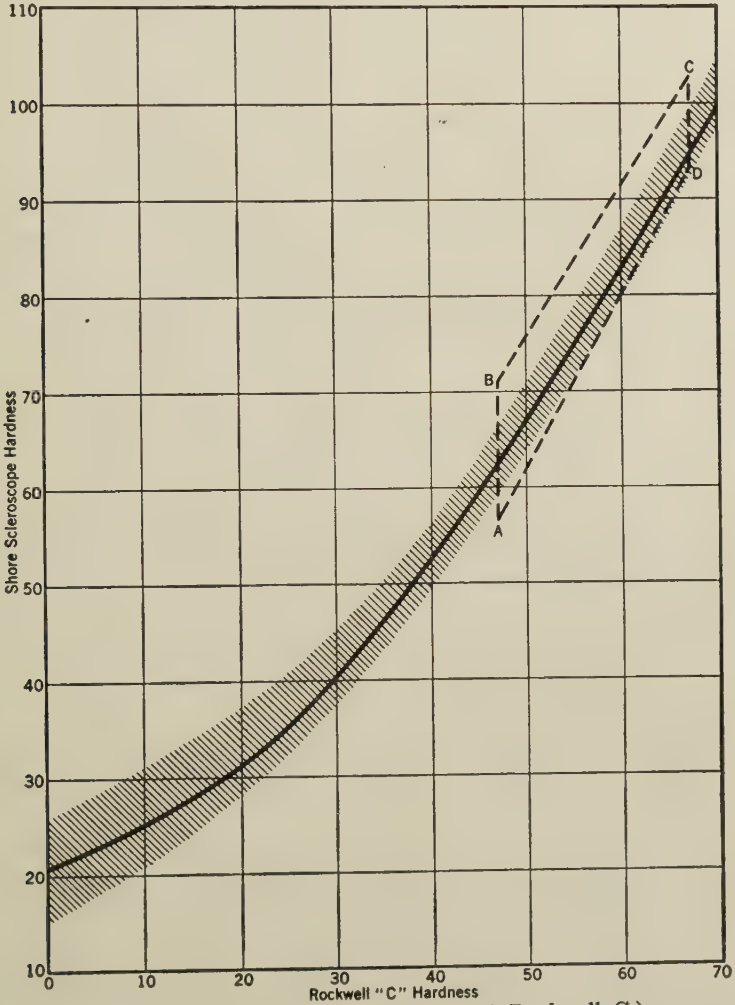

FIG. 5.—Conversion chart. (Shore and Rockwell C.)

Figure 6, Rockwell B to Brinell, shows fair convertibility in the range 120 to 220 Brinell.

All these figures show that if the hardness of two steels is to be compared, particularly two hard steels, the measurements are much more comparable if made on the same machine than on two different machines through conversion tables or formulae.

Gagnon[32] emphasizes this difficulty of making precise conversions.

Hardness *vs.* Tensile Strength. The indentation hardness values bear a rough relation to the ultimate tensile strength and may hence be used as a rough and ready approximation to the tensile values, but not as a sure substitute for a tensile test. Figure 7 shows that reasonably good conversion, within about 10,000 lb., in.,² can ordinarily be had, between 10 and 50 Rockwell C. Up to around 600 Brinell, where the Brinell values become gravely in doubt, the Brinell multiplied by 500 gives a fair approximation of the tensile strength, though this conversion is a trifle low below about 200 Brinell. Figure 8 shows the average of a number of conversion tables. The approximation—500 × Brinell = Tensile—is ordinarily within the limits of accuracy of the tensile and hardness tests. In plots of properties in this book, when tensile strength is plotted, hardness will not be shown, since it is only necessary to divide the tensile by 500 to get the Brinell to a degree of accuracy that is ordinarily sufficient.

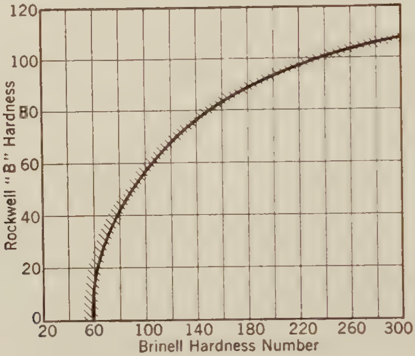

FIG. 6.—Conversion chart. (Brinell and Rockwell B.)

Vanick and Eash[130] presented conversion curves among Brinell, Rockwell, Vickers and Shore for very hard steels, and commented on the differences between results obtained from German and American scleroscopes, by which the German data give higher hardness numbers. Their curves and comments should be consulted in connection with the problem of hardness conversions.

A convenient conversion table covering Brinell, with steel and tungsten carbide balls, all the Rockwell and superficial Rockwell scales, Durometer, Monotron, Duroscope, Scleroscope and Herbert hardness tests has been published by Clapp and Clark.[131]

Tensile Testing. The primary means of evaluating the strength of a piece of steel, and, hence, its suitability for service, is the tensile test. In this is determined the load the piece will carry under a steady pull before actually breaking. During the course of the test, there is also determined the load *below* which, if the load were released, the specimen would return elastically to practically its original dimensions and *above* which, on continued loading, it begins to yield and become permanently deformed, or, more commonly, the load at which a tiny, predetermined amount of permanent deformation occurs. After the bar is broken, the amount it has stretched in length and the amount it has contracted in cross section are measured.

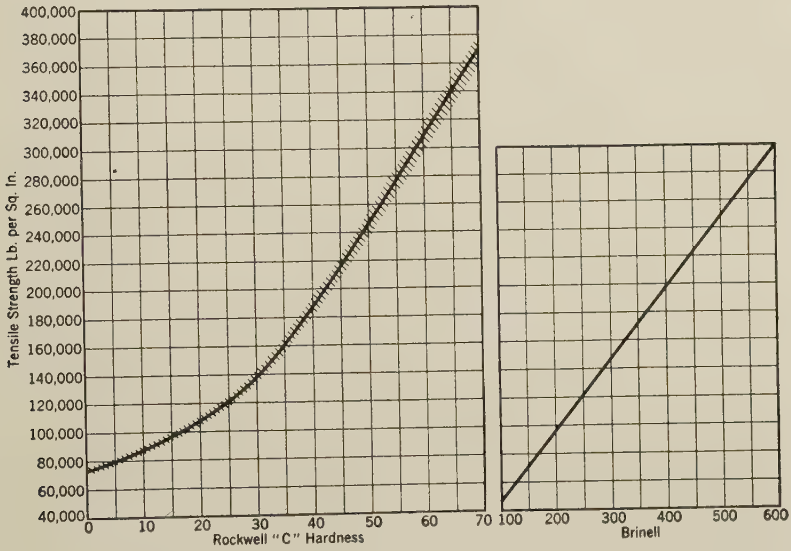

FIG. 7. FIG. 8.

Conversion charts. Tensile Strength. (Rockwell C and Brinell)

These loads in pounds, divided by the original cross section of the bar in square inches, give the stresses in pounds per square inch corresponding to complete failure by fracture, i.e., tensile strength, or ultimate tensile strength; and to elastic failure, i.e., elastic limit, proportional limit, yield point, or yield strength. The differentiation among these terms will be discussed below.

It is natural that the elastic strength should be calculated on the original cross section of the bar. For convenience, the tensile strength is similarly calculated. The true stress carried by the bar as the load increases from the yield stress to the ultimate while the bar is being pulled down to smaller cross section gives a useful picture of the behavior of the specimen during plastic deformation, but continuous measurement of the diameter during the test is so difficult that the true stress is seldom determined. The load at fracture divided by the reduced area after fracture may be calculated from the data regularly determined. This shows the maximum strength of the material after severe cold work, but the figure is seldom included in the list of properties.

Abbreviations used in reporting tensile and yield strengths, to avoid repetition of "lb./in.^{2}," are "p.s.i.," the Greek letter ψ (psi), or merely "tensile," with "lb./in.^{2}" implied. As stated in the preface, the last-named, letting "lb./in.^{2}" be implied, will be used in this book.

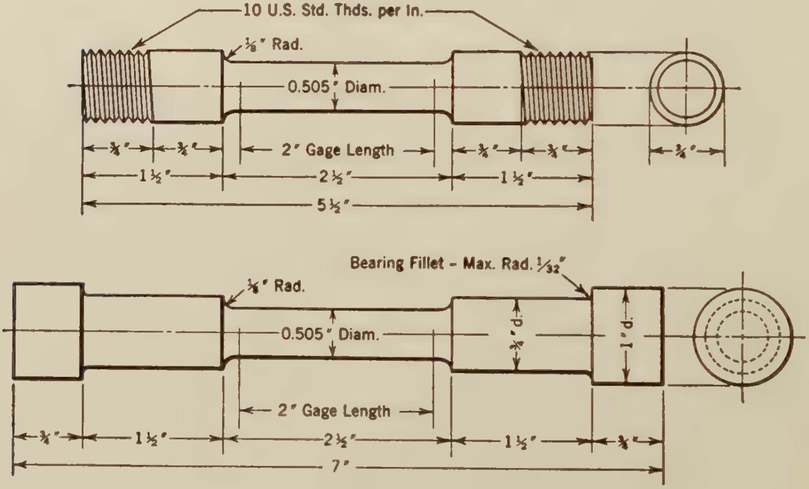

Fig. 9.—Standard tensile test bars.

The term "kip" is sometimes used to denote a unit of 1000 lb., thus 100,000 tensile might be reported as 100 kips/in.^{2} The British report strength in long tons (2240 lb.) per square inch and Continental observers in kilograms per square millimeter. The relation for conversion is 1 kg./mm.^{2} = 1422.32 lb./in.^{2}, usually rounded off to 1420.

Tensile Test Bars. The standard round bar used for most heat-treated steels is, in the United States, 2 in. between gage marks and 0.505 in. diameter. This diameter corresponds to a cross section of 0.2 sq. in., thus facilitating computation. The British bar is 0.564 in. diameter with 2 in. gage length, the German either 1 cm. diameter and 10 cm. gage length, or 2 cm. × 20 cm. The French is 1.38 cm × 10 cm. The standard U. S. rectangular bar for soft structural steel plate and the like is 8 in. between gage marks.

Figure 9 shows the standard 0.505 in. test specimen. Accepted dimensions for bars cut from pieces too small to give a standard bar are shown in the "A.S.M. Handbook."[33] By use of suitable holders bars without threaded ends may be used, provided proper precautions are taken for axial loading. Most tensile data in the literature are on specimens cut from 1-in. round bars. When heat-treated steels are tested at high hardness, the bar is usually machined to about 1/16 in. over the 0.505 diameter before treating and finally turned or ground to size. If the steel does not harden throughout, the results on bars of different diameters will be widely different. The section of the test bar should be comparable to the section in which the material is to be used. In evaluation of steel to be used in very heavy section it is customary to test bars taken from the center, the midway point, and the edge, so as to survey the whole section.

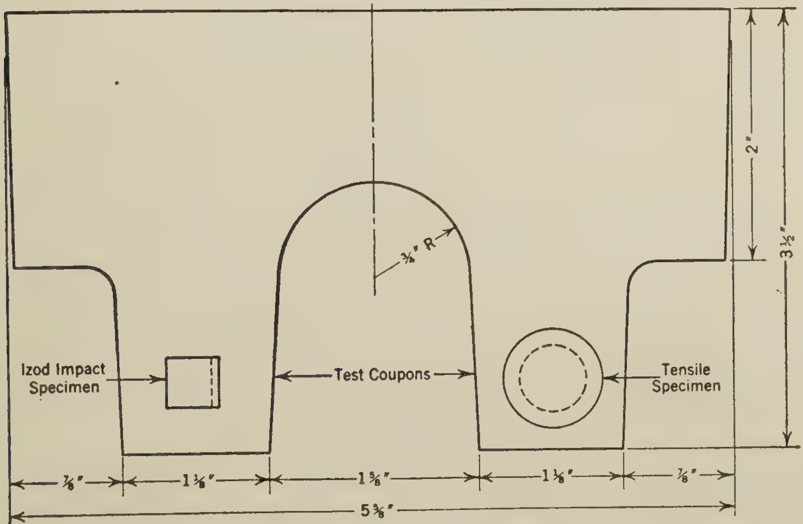

Fig. 10.—Cross section of keel block for cast steel. The length may be as desired. A length of 14½ in. gives stock for two tensile test bars from a single keel. The feeding head may be made taller if desired, the height shown is the minimum.

The usual specimen for cast steel is made in the form of a "keel" block with one or more projections about 1 in. wide, 1 to 2 in. high, and of length sufficient for one or two standard bars. Above the projections is a large mass of metal which is provided to ensure proper feeding and avoidance of shrinkage. The keel block is shown in Fig. 10 and is fully described in the American Foundrymen's Association "Cast Metals Handbook."[34]

It should be remembered that the 0.505 bar cut from a 1-in. thick keel does not have the same rate of freezing as a heavier or a lighter section of an actual steel casting, also that the thorough feeding of the test bar may not be paralleled in every part of the casting.

Considerable care needs to be taken in interpreting the results of any tensile test into the strength of a structure using the steel evaluated by the test, since the specimen may not exactly represent the material in the structure. The purpose of the text is more often to establish the suitability of the material than the properties of the structure. This is especially true of cast steel where, because of difficulties of feeding, all parts of the casting will seldom be as sound and dense as the test bar.

Yield Strength. The method of determining the elastic limit or the yield strength needs to be clearly stated. The elastic limit is the stress at which the bar will just fail to show permanent elongation, i.e., that value below which, if the load were released, the bar would return to its original length; thus showing all elastic and no plastic behavior. It is possible to determine this value by applying and releasing progressively higher loads, remeasuring the length of the bar. When the stress above which the first perceptible permanent deformation occurs is so determined, the result is the "elastic limit," and the yield strength was formerly so denoted, though seldom actually determined in that way. As a "go" or "no go" test, a "proof load" equal to the specified elastic limit is sometimes applied, any permanent stretch denoting failure; but when no permanent stretch is found one does not know how much higher a load could have been endured.

Releasing the load slows down testing; but a determination can be made by noting the extension for each increment of load, and plotting, strain against stress or getting the plot on an automatically recording testing machine. As long as the bar is behaving in a truly elastic fashion, stress is proportional to strain, and the plot will be a straight line whose slope is the modulus of elasticity. When plastic deformation starts, the line of the plot will begin to curve. The stress at the point of tangency of the curve and the straight line is the "proportional elastic limit." Unfortunately, the location of this point depends upon the sensitivity of the measuring device used—the more sensitive it is, the lower the proportional limit.

Unless the sensitivity of measurement is very definitely specified and adhered to, different observers will report rather variant figures for proportional limit. The true proportional limit is lower than the practical load-carrying ability for most conditions of service in which a very tiny permanent deformation can be endured.

By slightly overstepping the yield strength in a soft, ductile steel, the rate of elongation at a steady rate of loading is increased so fast that the beam of a lever-type testing machine will drop, the extensometer show a marked increase in rate of motion, or a pair of dividers held in the gage marks will tend to jump out. This crude "yield point by drop of beam" has been much used in rapid, rough-and-ready, testing but is not applicable to many of the strong, heat-treated steels whose stress-strain curves merge gradually from the straight line into a curve without the very sudden extension of the softer steels. Figure 11 shows typical stress-strain diagrams for a soft and a hardened steel.

A method is needed which will compare the elastic properties of all steels, both the very soft and the very hard. This is attained by setting up a definite amount of permanent elongation as the criterion of yield strength. This is generally either 0.1 or 0.2% of the 2-in. gage length. The report should state which one is used; 0.2% is more common and is implied when not stated. This specified set is laid off on a plot of the stress-strain diagram as an intercept and a line with that intercept drawn parallel to the beginning of the stress-strain line for the steel. This line cuts the stress-strain curve at the yield strength. Different observers can readily get duplicate values for yield determined in this way; so the method is coming into general use and the term yield strength is supplanting the older terms elastic limit, yield point, and proportional limit. Figure 12 illustrates the application of this method.

Ductility. The deformation produced in fracturing the bar is, of course, a measure of its ability to be deformed, i.e., its ductility.* Measurements are made of the change in length and the change in cross section.

Elongation and reduction in area (contraction) are reported in percentages of the original dimensions, and will be herein abbreviated to % elong. and % R.A. The % elong. depends upon the dimensions of the bar as well as on the properties of the steel.

Effect of Dimensions. Tensile and yield strengths determined on the different standard bars are closely intercomparable.

The elongation of the same steel tested in different dimensions, such as those in vogue in different countries, is not exactly the same, being affected by the ratio of length to diameter. Janitsky [35] has sug-

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* Ductility is a rather general term. High ductility usually connotes formability and toughness, but other, more direct, methods such as bend and impact tests are used to supplement the ductility figures when these properties are to be evaluated.

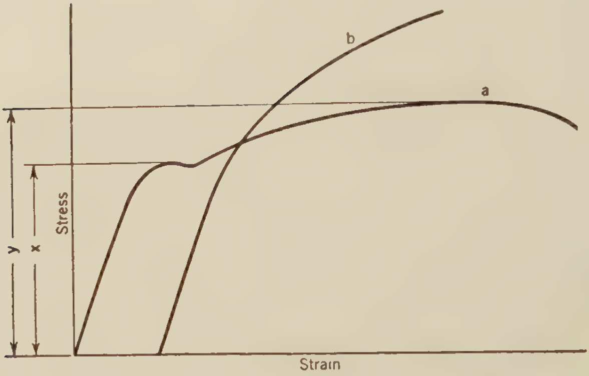


FIG. 11.—Typical stress-strain diagrams.

a — for soft steel
b — for hardened steel
x = yield strength for curve a
y = tensile strength for curve a
$\frac{x}{y}$ = yield ratio for curve a

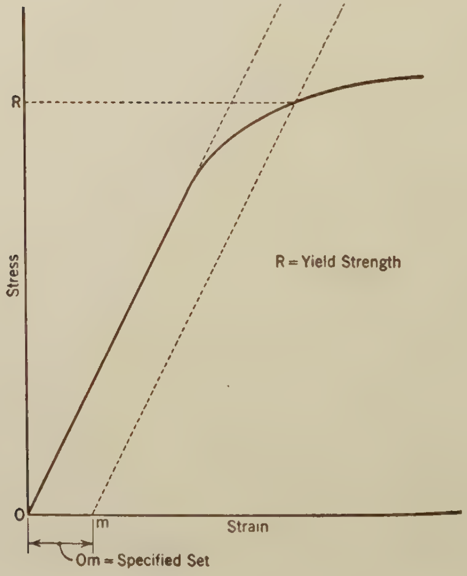


FIG. 12.—Intercept method of determining yield strength. (After A.S.T.M.)

gested methods for calculating what the elongation would be on a bar of one size when another size is used. More data than are usually reported are necessary to make a useful conversion when steels are compared whose tendencies toward general elongation and local necking are widely different. Reduction of area, at least when the necking down occurs near the middle of the bar, is not affected by the length; so this percentage figure is comparable for all round bars. Elongation values given herein are on the 2-in. gage length unless otherwise stated.

General Precautions in Tensile Testing. For accuracy and reproducibility in tensile testing, certain precautions are necessary. The testing machine must be properly calibrated and properly sensitive to changes in load. The specimen must be of standard dimensions (or, if a standard specimen cannot be taken from the material to be tested, any deviation reported), the specimen must be gripped so that it is loaded with reasonable axiality, and the rate of application of the load must be held within limits.

The requirements as to testing machines and their calibration and those on methods of tension testing are authoritatively set forth by the American Society for Testing Materials.[36]

Axiality of loading, in a machine that is itself properly aligned, is usually obtained by the use of threaded grips. Specimens too hard to be threaded can be made with shouldered ends and held in fixtures with spherical seats or ball-bearing alignment devices. Wedge grips, which serve on very soft and ductile materials, are not suitable for most heat-treated products.

The speed of pulling of the specimen, up to the yield stress, must not exceed 1/8 in. per min. Faster pulling leads to erroneously high results. Even the 1/8 in. per min. is far too fast for some special alloys. The testing temperature is not ordinarily specified, since it is expected to lie within the normal room temperature range and the effect of such variations is ordinarily not great. It cannot be assumed without proof that this is the case for all steels, and some observers think that R.A. values tend to be low in cold weather. A slight tendency in this direction is shown by data reported by Kinzel, Crafts, and Egan.[37]

Modulus of Elasticity. In the plot of a tensile stress-strain diagram, the modulus of elasticity is shown by the slope of the line below the elastic limit. The modulus enters into many engineering calculations, which generally take it as 29,000,000 lb./in.^{2}.

Few steels have moduli that deviate much from this value, and the modulus cannot be materially altered by heat treatment. When it becomes necessary to determine the modulus with exactness, test bars of long gage lengths are used and precision extensometers employed. Routine tensile tests on standard specimens do not give sufficiently accurate data for such a determination.

Engineering Application of the Tensile Test. As a criterion for load-carrying ability under static stress the yield strength is paramount. Few structures are useful after they begin to stretch permanently, and most engineering design is either avowedly or actually based on yield strength as the upper possible limit of loading with, of course, some factor of safety whose magnitude is set by the type of service. Therefore, much alloying of steel and much heat treatment are done with the primary purpose of improving the yield strength.

The designer is also concerned with how the steel acts under an overload for which it was not designed and with its ability to distribute localized stress. He desires to avoid brittle failure and give the steel the ability to "hang on" and take punishment before it finally breaks. Hence, although he wants a high yield strength he does not want it at too great an expense in toughness. A high yield ratio, i.e., proportion of yield strength to tensile strength (See Fig. 11), is acceptable if there is sufficient toughness in the interval between the yield and the tensile failure. In spring steel a compromise is made. The maximum possible yield strength is not sought, but, nevertheless, some toughness is sacrificed in favor of high yield strength, because both cannot readily be simultaneously obtained in high degree.

If the tensile strength is very considerably greater than the yield strength, there is likely to be ample toughness, but this is not necessarily true. Poor ductility may accompany a low yield ratio. What is usually wanted is the highest yield strength compatible with great ability to absorb energy beyond the yield. One criterion for judging this ability is the area under the stress-strain diagram, and a complete stress-strain diagram is one of the most satisfactory methods of evaluating the static properties. It is inconvenient to determine this beyond the yield; so the evaluation is more commonly made on the elongation and contraction measured after fracture.

To the experienced eye, the appearance of the fracture, i.e., whether it is "flat," "cup and cone," or "star," is an indicator of the degree of ductile flow occurring beyond the yield stress.

Ductility Requirements. Much discussion has arisen between engineers and metallurgists as to the amount of ductility required in various types of service. The metallurgist agrees that there should be sufficient ductility to ensure avoidance of brittle failure, but he questions the tendency of the engineer to demand, along with higher yield strength, the same elongation he has been accustomed to getting in weaker materials, irrespective of the ability to absorb energy. Bridge engineers are especially conservative in this respect. Some engineers admit that they can conceive of no service in which more than 10% elong. can be usefully employed, but they are in the minority. In most cases, rather high elongations are specified.

The vital importance of high elongation in heat-treated steel is questioned by many metallurgists and engineers, and more emphasis is placed on the reduction of area in association with the character of the fracture. High reduction of area and a "cup and cone," or "star," fracture are taken as an indication of the ability of the steel to withstand deformation without breaking; and it is this factor which engineers have been trying to incorporate in their specifications. This is made evident by the fact that in past years, when the heat treatment of steel for structural purposes was in its infancy, and forged and rolled material was either untreated or simply annealed, the bend test was required in addition to the percentage elongation. Experience has shown that a minimum R.A.—probably about 35% for straight C steels—is necessary to meet the usual 180° bend requirement, regardless of the fact that the percentage elongation may be good. The yield strengths and elongations of two similar steels may be alike and yet one may fail to meet the bend test. This is illustrated by the following tests from heavy forgings, two annealed and two heat treated, yet all four from the same heat of steel.

TEST	ANNEALED		HEAT TREATED	
	(*a*)	(*b*)	(*c*)	(*d*)
Tensile	70,000	71,250	85,750	88,500
Yield	37,500	37,000	51,750	52,000
Elong. %	25	25	23.5	24
R.A. %	33	47.5	32.5	53
Bend test, 180°, over a pin whose diameter equals the thickness of the specimen	N.G.	O.K.	N.G.	O.K.
Fracture	angular	cup	granular	silky

The argument for high ductility is based not solely on the actual need for such toughness but partly on the viewpoint that when the steel has been melted, cast, fabricated, and heat treated in the best commercially practical manner it can be made to show the combination of properties demanded by the specification, and anything less indicates something wrong with the steel which might be manifested in service in ways not directly shown by the tensile test. Thus, the combination of tensile test properties is specified in order to ensure that the material shall duplicate that known to have given good results in service, without too definite relation to the ductility from the point of view of actual engineering need for it.

The National Advisory Committee for Aeronautics points out[38] that "for some metals the values for elongation and reduction of area in a tensile test are required to be two or three times as great as in others," and the committee is examining the relation of these values to service requirements.

The engineering justification for a particular ductility required by a specification needs to be considered separately from the matter of ductility as an index of quality and uniformity.

The automotive industry is notably progressive in trying to evaluate just what properties a given service requires. For example, during the war, Liberty motor crankshafts were made from highly alloyed steel forgings in which great ductility was demanded and secured. For a long time similar toughness was demanded in automobile crankshafts, even while they were being made larger and more rigid. Finally, the Ford cast crankshaft was evolved, which shows almost no elongation at all and very low impact resistance.[39] Similarly, Almen and Boegehold[40] state that, since a rear axle gear is not called upon to resist impact, they do not exclude steels of good wear resistance but low impact for this service.

Whatever the future holds as an ultimate outcome of the ductility problem, the engineers' requirements will doubtless continue to be expressed primarily in the terms of the tensile test.

Directional Ductility. Since so much steel is used in bar form, from which standard tensile specimens can be taken only in the direction of rolling, the tensile properties usually reported are those from such longitudinal bars. Specimens of plate stock can be taken longitudinally or transversely, and, on prolongs of large forgings, longitudinally, transversely, or radially. When specimens are tested from plates that have not had much cross rolling, or from forgings not uniformly worked in all directions, the tensile and yield strengths taken in other directions are normally very close to the longitudinal figures, but the elongation and reduction may be shockingly low. In gun forgings, the working pressure is applied radially, so arsenals evaluate the forgings with very great regard to radial ductility.

These directional properties are due to the fibering-out in working of the original dendritic structure of the casting, of inhomogeneities that produce banding, and of non-metallic inclusions. Steel very free from non-metallic inclusions, heat treated to break up the dendritic structure, and uniformly worked in all directions will show transverse and radial ductilities equal to those determined longitudinally, and, where service stresses are to be applied other than in the direction of working, the ratio of directional ductility tells much about the suitability of the steel. It is because of differences in directional properties that such great pains are taken to secure the proper direction of the grain in upsetting forgings for gear blanks.

Unfortunately, one cannot always work the steel equally in all directions, and in such cases transverse and radial ductility figures are not as high as those on longitudinal specimens.

Shearing Strength (Torsional Strength). The shearing strength of steel is of significance in considering the strength of shafts and axles, which are subjected to torsion, and in considering the strength of bolts and rivets, which often are subjected to direct shear. The torsion test is the test best adapted to determine elastic strength in shear; and by the use of hollow specimens of suitable design the ultimate shearing strength of steel can be determined. It is doubtful whether the commercial determination of shearing strength will become common, because for most steels the shearing elastic strength lies between 50 and 60% of the yield strength; hence the yield strength in tension serves fairly well as an index of shearing strength. The ultimate shearing strength of steel usually is somewhat more than 60% of the tensile strength.

Compressive Strength. The compressive strength of steel is best determined by testing small cylinders with a height equal to two or three times the diameter. Such cylinders must have their ends machined very carefully to plane and parallel surfaces, must be centered very carefully in the testing machine, and load should be applied very slowly through spherical-seated bearing blocks.

Only for very brittle steels is it possible to determine the ultimate compressive strength. For all ductile steels the specimen flattens out under load, and there is no well-marked fracture. The elastic strength of steel in compression can be determined by using apparatus and methods similar to those used for determining elastic strength in tension. The compression test is not much used as a commercial test for steel, because most available test data indicate that the elastic strength of structural steels in compression is not widely different from the elastic strength in tension; hence the tensile test serves fairly well as an index of compressive strength. However, the compression test may become an important test for structural steel for columns, since this directly determines the property utilized in service.

High-temperature Testing. Engineering advances, particularly those in the cracking of petroleum, in the generation of steam power at higher temperatures and pressures, and in some chemical industries, have brought demands for special steels with suitable heat treatments to serve the elevated temperature requirements. The temperatures in question range from, say, 400° to 1400°, and in some cases to 2000°. The plastic flow met above the yield stress in room temperature testing begins at much lower stresses at high temperatures and may be but very slowly developed. The room temperature properties are no criteria for properties at high temperatures, and even a tensile test at the temperature of operation fails to reveal the full tendency toward plastic flow, since this tendency appears so slowly. The regular tensile test carried out at high temperatures, of course, gives useful information as to resistance to sudden overload at such temperatures, but it has little relation to the stresses allowable in design. Even a very tedious proportional limit determination with highly sensitive instruments, made in the course of the high-temperature tensile test, is of extremely doubtful value.

Creep Testing. Hence, very long-time continued loading tests are required when stability of dimensions must be assured, the long time of loading being used not so much to allow the accumulation of sufficient deformation at reasonable loads to be accurately determinable as to give time for a fairly stable rate of deformation to manifest itself. Such tests, termed "creep" or "flow" tests, are normally carried out for at least 1000 or 1500 hr. when, if the rate of extension is holding constant, the results are extrapolated to the extension that would apparently be met in 10,000 or 100,000 hr.

Even 1000-hr. tests involve grave uncertainties in extrapolation to service lives, but they have been found useful in the comparison of materials for high-temperature service. Such comparisons are often made on the basis of the stresses producing a calculated creep of 1% in 10,000, or 100,000 hr.

Both short-time, high-temperature, tensile testing and creep testing require extreme attention to holding the temperature constant and to making sure that the whole gage length is at the desired temperature, since in some ranges of temperature the load-carrying ability falls off with amazing rapidity as the temperature rises. The very special precautions necessary in these types of testing have been summarized.[15]

High-temperature tests of 50 or 100 hr. duration are in some vogue abroad, under the assumption that by comparing the early stages of the curve so obtained with those of a previously made long-time creep curve for a supposedly similar steel it may be taken that curves coinciding in the early stages will coincide thereafter. Much evidence, however, is at hand to show that this is not necessarily true; so the "accelerated" tests find few adherents in the United States and are not approved by the National Physical Laboratory of England. There is need for a reliable rapid acceptance test and many have been proposed, but proof of the reliability of any is still lacking. The creep problem will be discussed in Vol. II, Chap. 19.

Fatigue. Since engineering design has been based primarily upon yield strength or tensile strength, with some factor of safety, few machines fail from static overloading unless they are involved in an accident which throws upon them a stress far exceeding the design value. However, many failures do occur in parts subject to repeated stress well below the yield stress that can be borne in a single static loading. Indeed, it is often stated that 90% of actual fractures occur owing to repeated stress; and drivers of automobiles of the vintage of a dozen years ago will remember cases of failures of springs and axles, probably in their own experience, when the part broke off with no sign of deformation, with a fine-grained "detail" fracture over most of the break, and with signs of the ends of the break having rubbed together for a long time before final failure. Perhaps the final fracture was coarse grained in comparison to the detail fracture, from which arose the garage mechanic's erroneous phraseology that the part had "crystallized in service." Such fractures start from a nick or a notch of some sort (external or internal, perhaps a non-metallic inclusion in the latter case), or a poor fillet that brings about a concentration of stress at some local point. Under long-continued repetition of stress, a crack finally starts at this locally overstressed point and then progresses to final failure.

Much attention has been paid to the problem of fatigue failure, and its nature is now quite well understood. At stresses higher than a certain stress, termed the endurance limit, there is a tendency for the steel to be locally overstressed (but without material plastic flow, so that the overall dimensions are not affected.) Though it is difficult to prove, it is not unreasonable to assume that this spot, because of some local defect of the notch type, or because of inability of the material to distribute the stress, is stressed above its yield strength and might be considered to become finally so hardened by cold work that it becomes brittle and starts to crack. The damaged spot may be so submicroscopically small that no evidence of its existence can be obtained till the crack actually starts.

In contrast to this damaging effect of overstressing, there may be a helpful effect of understressing. Stresses below the endurance limit tend to produce something analogous to cold working, again without perceptible changes in overall dimensions, but with a change that is reflected in a perceptible, though small, increase in hardness and tensile strength.

These two effects, damage by overstress and improvement by understress, come to a balance at an intermediate stress, the endurance limit. This is the maximum stress which the specimen can endure for any number of repetitions, however large.

The endurance limit is of very great importance in design of repeatedly stressed parts. Only in very soft steels is the endurance limit higher than the yield strength. Some such soft steels can be pulled out and pushed back by rapidly applied tensile and compressive stresses that are perfectly balanced, as in a rotating shaft, without showing either permanent deformation or failure from fatigue, though if the same range of stress were applied with the stresses unbalanced, deformation would ensue.

In any but the softest steels the endurance limit is below the yield stress and bears no direct relation to it. Often a steel will fail under repeated application of loads that are entirely safe from the point of view of static loading. This accounts for the service failures from fatigue when design pays no attention to the endurance limit.

The demands of design made it necessary to evolve methods of fatigue testing. The most convenient is the method of rotary bending; i.e., a shaft is loaded and rotated so that each element of the surface is alternately loaded equally in tension and in compression. Axial loading is also used, when the tension and compression loads may not be equal. However, the average of these unequal loads has about the same effect as that of fully balanced loads equal to this average, as long as the stress range is not high enough at either end to exceed the yield strength in tension or compression. The main difference in the two types of loading is that axial loading stresses the whole cross section uniformly, whereas rotary bending has the maximum stress at the surface. Hence case-hardened or nitrided steels or those that have been work-hardened on the surface, i.e., those with fatigue-resistant surfaces, show higher endurance limits in rotary bending than in axial loading, and the results of rotary bending are not directly applicable to the performance of non-homogenous steels unless they are thus stressed in service.

Ordinarily, however, the rotary test is adequate to evaluate the endurance properties of a steel. The test is carried out on several types of specimens and machines, machines of the R. R. Moore type being now most used.[41]

In making the test, a series, say not less than four, of like specimens, carefully polished and free from transverse scratches, with the breaking section having a radius of ·1 in. or more, is prepared, and the first one is run at such a high load that it fails in a relatively small number of cycles, say a few hundred thousand. New specimens are run, each at a slightly lower load, till a load is found at which the specimen remains unbroken for at least 5 and preferably 10 million cycles. Experience has shown that in ferrous materials a specimen that will run 10 million cycles unbroken will, with only rare exceptions, continue to run for an indefinitely large number of cycles. This is not necessarily, and not usually, true for nonferrous materials which should be run to 50 or 100 million cycles and the results reported as "endurance at so many cycles."

Failure may occur after running more than 10 million cycles in very hard steels,[42] and the endurance limit for springs, especially those made from wire, is generally very far below the value one would expect from its tensile strength.[43, 44, 45, 46] The endurance ratio may fall below 0.30. Such behavior is generally ascribed to the presence of internal stress, incipient cracks due to overdrawing, decarburized surfaces, and to the greater effect of surface scratches on very hard steels.

For steels, in the ordinary condition however, the 10 million cycle figure seems justifiably reported as a true endurance limit. Even so, the test is time consuming and not well adapted for an acceptance test. There is no known satisfactory means of accelerated fatigue testing, though it has long been sought.

The best way to accelerate the test is to build the testing machine in perfect alignment so that it will run smoothly at high speed, and use very high r.p.m., e.g., 10,000, which is now quite common. Another short cut is to run, say, to 5 million cycles only, and then raise the stress, for example, to that corresponding to ¼ million cycles on a virgin bar. A damaged specimen will have a somewhat shorter life. If the specimen was strengthened and not damaged in the original test, it will run *more* than ¼ million, perhaps 1 or 2 million, and this justifies the assumption that it would have run indefinitely at the original stress, the evidence being at least as good as if it had originally been run to 10 million. (Compare the dotted lines in Fig. 17.) The total testing time is shortened by 3 or 4 million cycles.

The number of cycles to fracture at each stress is plotted against the stress, usually on a semi-logarithmic scale, as shown in Fig. 13. Thus, stress vs. number of cycles curves are generally abbreviated to "S-N curves." The stress at the value where the curve becomes horizontal is taken as the endurance limit.

If a steel is truly uniform, as the steel in Fig. 13, duplicate runs will give an S-N curve with little scatter, whereas if there is a lack of uniformity the curve will be a wide band. S-N curves would be a very useful evaluation of uniformity were it not so tedious to obtain them.

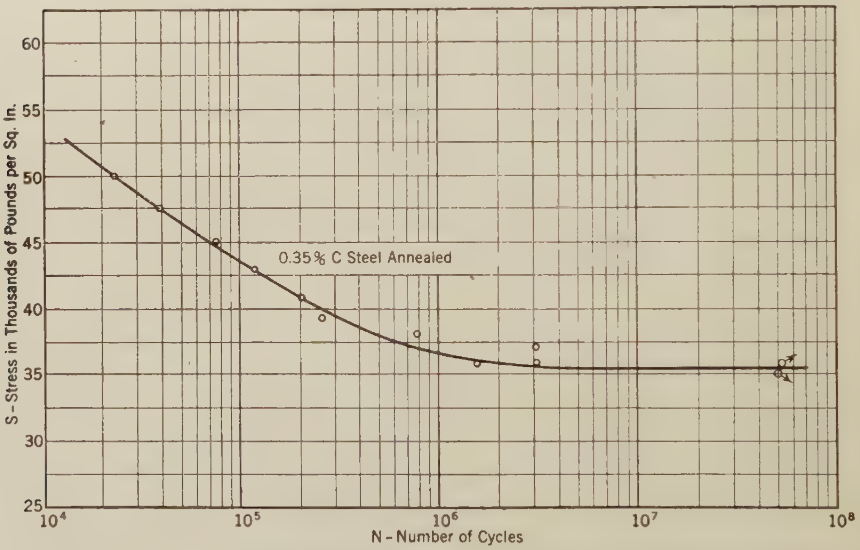


FIG. 13.—S–N curve for determination of endurance limit. Note the logarithmic plotting for cycles. This is a semi-log plot. Log-log plots, with stress as well as cycles plotted on logarithmic scales, are also used. The arrows indicate specimens unbroken after 50 million cycles.

For no well-understood theoretical reason, it turns out that the endurance limit for steels and ferrous materials generally is ordinarily not very far from half the tensile strength. The endurance ratio approximates 0.50. Since the Brinell hardness also happens to have a nearly linear relation to tensile strength, it follows that Brinell hardness is more or less linearly related to endurance limit; if the Brinell value is multiplied by 250, the result approximates the endurance limit.

However, this empirical rule is only approximate, and the upper band in Fig. 14 shows how far from an accurate rule it is. Not only does the normal value drop below the ratio of 50% of the tensile strength above about 175,000 tensile, but, even below that strength, steels of the same tensile strength may vary 80% in endurance limit, even when the specimens are highly polished and free from surface scratches and notches. This deviation can seldom be traced directly to variations in the usual heat treatment of the steel though, of course, the improvement in tensile strength produced by heat treatment is ordinarily directly reflected in an analogous increase in endurance.

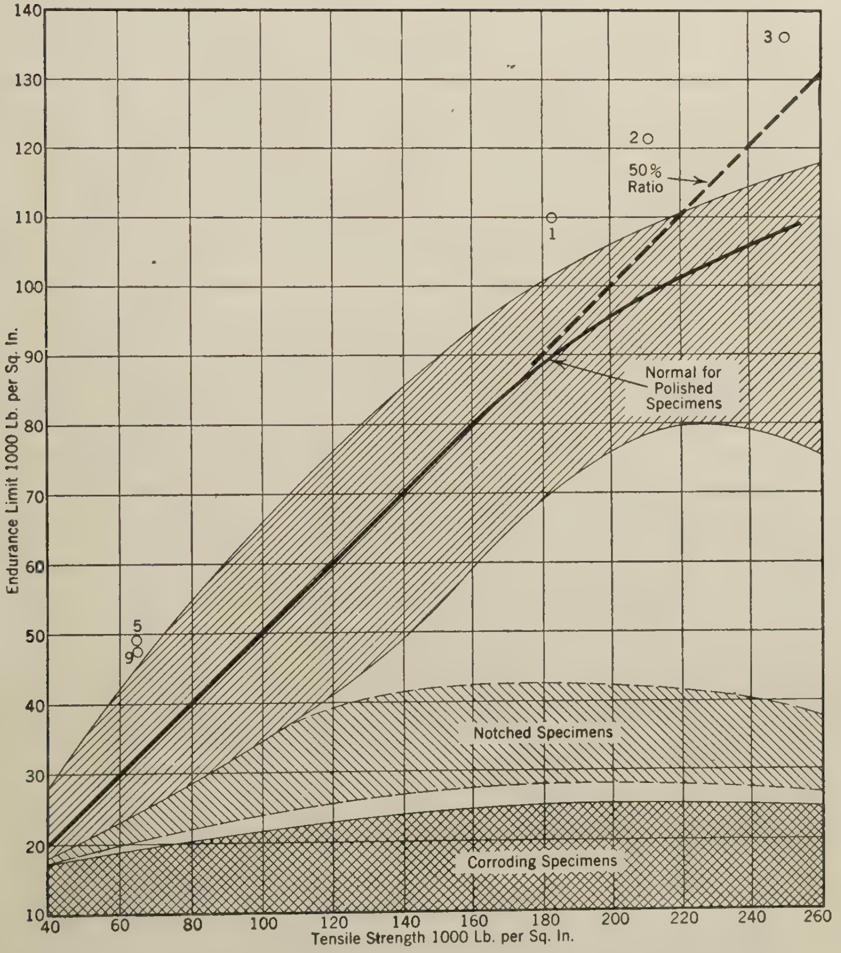


FIG. 14.—Relation between tensile strength and endurance limit.

However, isolated cases are on record of endurance limits for strong steels given heat treatments aimed to free the steel from internal stress or to prevent internal cracks, which indicate that it should be possible to obtain values that hug the upper boundary of the band for polished specimens in Fig. 14. Points marked (1) and (2) were obtained by long-continued tempering of oil-quenched Ni-Cr-Mo steels,[47] point marked (3) is for an "austempered" C steel[48] (see Chap. 5), whereas points (5) and (9) represent high-yield strength, low-alloy, non-heat-treated steels of high P content[49, 50] of the type to be described in Vol. II, Chap. 11.

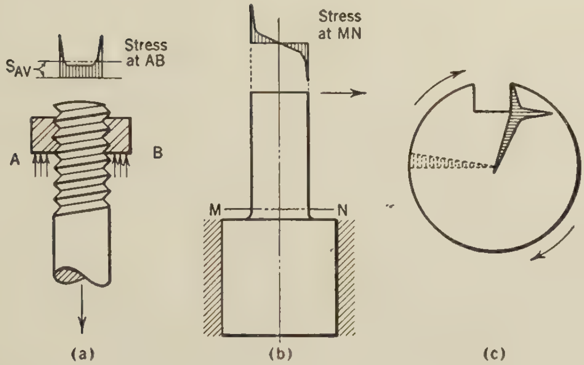
Stress Raisers. Not only do different steels and different lots of similar steels act differently in the standard endurance test with smoothly polished specimens, but they vary also in behavior when notches are intentionally or unintentionally produced so as to cause local increase in stress.

Even the most notch-resistant steel is severely and very adversely affected by notches. Most fatigue failures in actual parts are directly traceable to some local irregularity such as a tool mark, a number or letter stamped into the steel, a key way, or a fillet of too small radius. Figures 14 and 17 show how much the endurance is decreased by the presence of a sharp notch. Moore[51] has sketched a number of the common types of stress raisers and indicated the stress increase due to the geometric shape of the piece (Figure 15). Disregard of stress intensification is a sure way to invite fatigue failures.

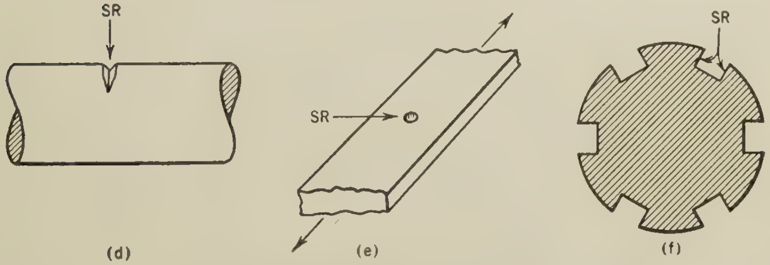
Determination of Stress Distribution. If a scale model of a steel part is made from a transparent material like glass or celluloid, loaded, and examined by polarized light, no color bands appear if the stress is distributed uniformly. Localization of stress produces colored bands whose number is a function of the stress and whose location shows the position of the stress intensification. By such means the evil effect of notches, poor fillets, etc., may be visually demonstrated. This method of study is useful in the design of objects in which the stress concentration cannot be readily or accurately calculated.

Examples of the use of this method are found in the articles by Frocht[52, 53] whereas Peterson and Wahl[54] link together stress concentration and behavior in fatigue, as do Buckwalter and Horger.[55]

Notches. The experienced steel treater has a good picture of what features of design are stress raisers, since objects designed with them, when they are quenched, localize the stresses set up in quenching and the work often cracks. This is still more troublesome because the steel is hard, and, as Fig. 16 after Karpov[56] shows, damage by notches is greatest in hard steels. This figure is drawn on the basis



Typical stress-raisers in machine and structural parts. (Moore[51])



Imposed stress-raisers. (Moore[51])

(g) Inherent stress-raisers. Shatter cracks in a new rail. (Quick, Bureau of Standards.)

Fig. 15.—Stress-raisers that lead to fatigue failures under overstress, such as screw threads, sharp fillets, key-ways, nicks, holes, sharp corners, cracks. Surface seams and large, sharp-cornered, non-metallic inclusions fall into the same category.

of percentage loss from the endurance shown by polished, un-notched bars in air.

Specimens with square notches, V-notches, or having changes in section without fillets, give endurance limits ordinarily lying within the band shown in Fig. 14 for notched specimens. Unless notches are avoided, little profit is gained by using very strong steels. Some austenitic steels which work harden rapidly, however, are able to withstand the effect of a notch. Russell and Welcker[57] report such behavior on 18 Cr, 8% Ni stainless steel. Cast iron, also, is less sensitive to notches than are the ordinary steels. Figure 14 is drawn to represent the ordinary types of steels.

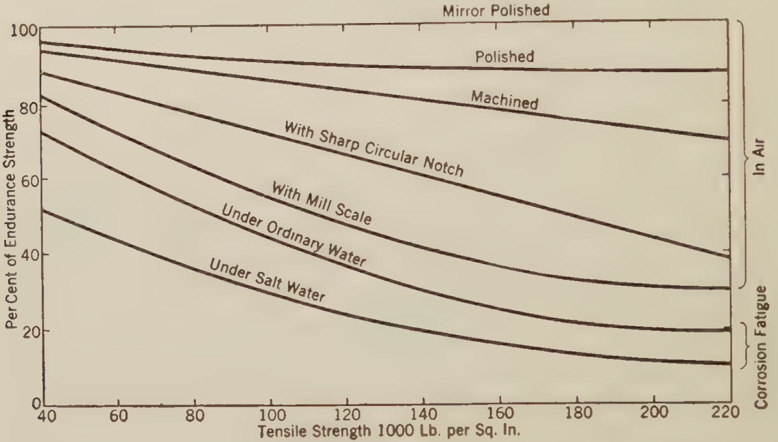


FIG. 16.—Reduction of endurance of steel depending on surface conditions. (After Karpov[56]).

Factors in Fatigue. The following points should be kept in mind:

1. Screw threads, grooves, sharp shoulders, nicks, and even tool marks cause high localized stresses and markedly lower the fatigue strength of any steel machine part.

2. In figuring allowable stresses under repeated stress it must be remembered that stresses above the yield strength of the metal will cause *elastic* failure. The machine designer and the user of steel parts must be on guard against both elastic failure and fatigue failure.

3. Fatigue strength obtained in tests of small specimens, especially of heat-treated steel, cannot often be obtained in large pieces of steel. It is as necessary to use a "factor of safety" in considering fatigue strength as it is in considering static strength.

But sometimes a part cannot be designed without severe stress intensification; so the ability of different steels or of the same steel differently heat treated to withstand the effects of stress raisers enters into the choice of the steels or treatments.

To perform well under stress-raising conditions, it is generally felt that high ability for local deformation, such as that evidenced by high reduction of area in the tensile test, is helpful in allowing the material to yield locally at the base of a notch and thus redistribute the stress and lower its intensity. However, the materials of high reduction of area are not always the best in notched fatigue tests. Not only is there required ability to redistribute stress, but also ability to resist the propagation of a crack once it is formed. The quality needed has been termed by H. F. Moore "crackless plasticity," [58] but how it is to be measured and evaluated is not yet certain.

In parts designed with reasonable attention to freedom from stress raisers [8] and to the endurance limit as determined on polished specimens, fatigue failures in service seem to come mainly from the imposition of a few accidental overloads; so the ability of a steel to withstand overload without damage counts perhaps as much as its endurance limit on a polished specimen. At any rate, some materials of mediocre endurance limit have good service records. They seem to have exceptional ability to stand overload without rapid damage.

Damage-line Testing. These factors have led, in comprehensive studies of fatigue resistance, to the addition to the already tedious standard endurance test of further endurance tests on notched bars, and on ordinary bars that have been subjected to a known number of cycles of known overstress and then retested at the endurance limit for virgin, non-overstressed bars, to see whether they have been damaged or whether they will run the usual 10 million cycles without evidence of damage in the prior overstress run. From such tests a "damage" line can be drawn, as French,[59] and as Russell and Welcker [57] have done. (See Fig. 17.) Such work is slowly throwing light on the properties needed to combat fatigue failures in service, though it is not yet possible to correlate them closely with properties determinable in static tests or in ordinary endurance tests.

The endurance of actual machine parts is not necessarily accurately evaluated by laboratory tests of polished or of notched specimens, as Almen and Boegehold [40] point out for rear axle gears, where it is necessary to test complete assemblies. But even in such cases present laboratory tests and the laboratory methods of attack do throw light on the problem, and study by the damage-line method may give further information.

In most types of service under repeated stress, the maximum stress applied is at the surface. If the surface skin is weak, fatigue failure is invited. The evil effects of a decarburized skin on a spring are common knowledge and have been shown in detail by Hankins and Becker[60] and by Austin.[61] Methods for hardening without decarburization are demanded, and the susceptibility to, or freedom from, decarburization is an important factor in the choice of an alloy steel.

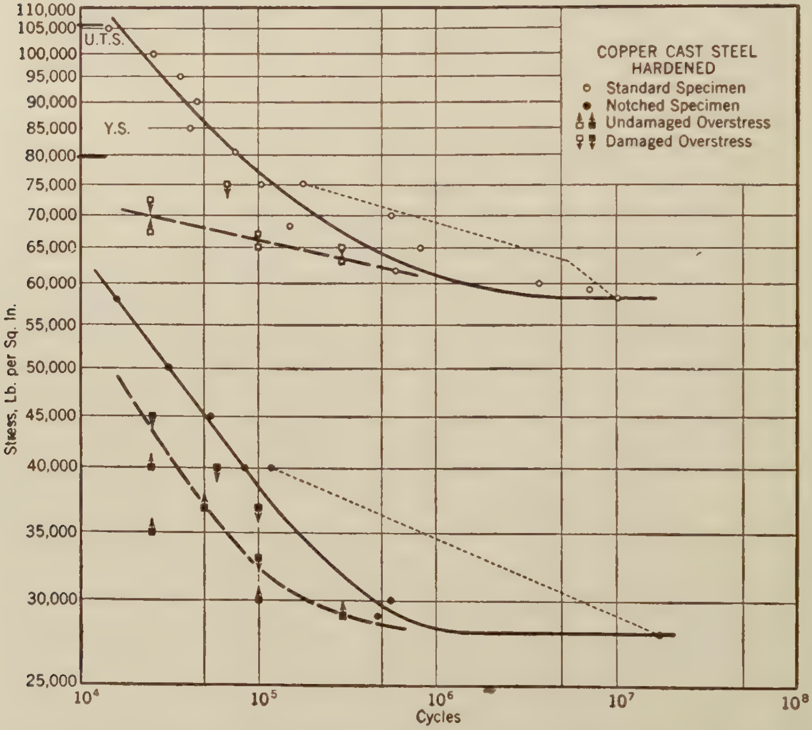


Fig. 17.—Log-log S–N curve for cast 0.23 C, 0.38 Si, 0.61 Mn, 1.20% Cu steel. 1650°-1 hr., air; 930°-3 hr. Tensile, 106,000; yield, 79,500; 24% elong., 48.5% R. A. Upper curve, polished specimens. Lower curve, 90° square (Kommers[125]) notch, stress calculated on diameter at base of notch. Dashed lines divide the fields for no damage (upon re-testing at the endurance limit of a virgin bar) by overstress, of intensity and number of cycles plotted, and for damage by such overstress. The dashed lines are "damage lines." The dotted lines connect the test of a virgin bar that was strengthened by understressing with its re-test at a higher stress. (Russell and Welcker[57]).

Conversely, if the surface is made stronger, as by carburizing or nitriding, the endurance will be raised in a rotary bending test or in service where the skin takes the maximum stress. Very great increases in endurance limit, for example, are produced by nitriding and small increases by cold working the surface.[62–65] However, if the stress is applied axially (as in a connecting rod, for example), this improvement, as shown by rotating beam tests, may be illusory since failure may start beneath the surface. A hard, but brittle surface layer in which a crack can readily be formed may be dangerous because the crack may readily propagate through the softer core. A case in point is the galvanized heat-treated bridge wire discussed in Vol. II, Chap. 1, and other cases have been described.[66]

Corrosion Fatigue. Finally, there must be mentioned the further fact that, since steels are sensitive to the effect of notches, the notches caused by corrosion are highly detrimental, and that, moreover, *simultaneous* repeated stress and corrosion are even worse than corrosion followed by repeated stress. The "corrosion-fatigue" phenomena, originally brought out by Haigh[67-68] and emphasized by McAdam,[69-71] have to be considered in any service involving the two factors. Sucker rods in oil-well service are an outstanding example.[72,73]

Under combined corrosion and repeated stress there is no true endurance limit, and the endurance of a very strong but corrodible steel is very little if any greater than that of a much weaker, equally corrodible steel. This is indicated in the lower part of Fig. 14 which shows that under active corrosive conditions the endurance of all corrodible steels falls practically to the same level. Under such conditions strengthening by heat treatment is unavailing.

There is no true corrosion-fatigue limit, the stress-cycle curve tends to be a straight line instead of flattening off, although in Figs. 14 and 16 the corrosion-fatigue results have been plotted as if they were endurance limits. If fatigue tests are carried out under corrosive conditions, a corrodible steel will ultimately fail, no matter how low the load, though failure may be long delayed. The behavior under combined corrosion and repeated stress indicated in Figs. 14 and 16 is for ordinary, corrodible steels. If an alloy steel is completely incorrodible under the corrosive conditions to which it is subjected, it will show the normal S-N curve, i.e., the same as tests in air. Even the most corrosion resistant steels are seldom so completely resistant as to show no loss of endurance in corrosion-fatigue tests.

Corrosion Testing. Corrosion by itself may not be a major problem in most services for which heat-treated steels are required, for the steels used in moving parts are generally automatically coated with oil, and steels subject only to static stress may usually be painted or otherwise protected. When the necessity does arise for corrosion testing, the important thing is to see that the test conditions duplicate, or are sufficiently analogous to, the conditions of service. No "standard" test for corrosion exists, because corrosion resistance is not a property inherent to the steel, exclusive of its service environment.[126-127]

Notched-bar Impact Testing. The discussion under fatigue has brought out the marked intensification and localization of stress at the base of a notch even when the applied stress is low. When a notched bar is broken by a single application of applied stress, very great differences appear between different steels, even when they show the same tensile and yield strengths and the same ductility in the tensile test. Some are tough and deform freely; others snap off almost like glass.

For instance, Sims[74] tells of two bars of cast steel from the same heat which were heat treated differently. These were tested by the tensile test and the notched-bar impact test. Though the results in the tensile test were nearly identical, they were materially different in the impact test, as the following data show.

	Heat Treatment A	Heat Treatment B
Tensile	98,250	101,500
Yield	63,000	66,000
Elong. %	27.0	27.5
R. A.%	55.5	57.5
Izod	18.5	48.0

Bar *A* was normalized from a rather high temperature after a long period of heating, resulting in a coarse-grained structure, bar *B* was normalized from just above the critical temperature, resulting in a fine-grained structure. One cannot say that one of the two treatments is right and the other wrong without knowing the type of service to which the steel is to be put. In high-temperature service the coarse-grained material with the lower impact resistance would be the better, whereas it would be the poorer in shock-resistant service at ordinary temperature.

The impact test shows *A* to be exhibiting relatively brittle behavior, *B* ductile behavior. These results do not tell us that steel *A* is always brittle and *B* always ductile. They only mean that these things are true in so far as the particular set of loading conditions imposed by the test is concerned. Some types of loading conditions could probably be found in which a steel with the properties of *B* would act no more ductile than one with the properties of *A*.

In other words, most steels have a certain propensity toward brittle behavior, which varies from steel to steel. This difference has some importance in relation to service involving sudden shocks of high intensity, and impact resistance may govern the choice of material in such service. However, many materials that are almost glass-brittle under impact in the presence of a sharp notch are tough when they are formed with fillets of reasonably large radius, such as good engineering design always calls for. Under the tensile-impact test, which uses an unnotched bar but applies the tensile stress by a sudden pull, they may show quite as good ductility as in a static tensile test; so the suddenness of the blow, up to ordinarily high rates of loading, is not what makes the steel act brittle.

On the other hand, at extremely high rates of loading, as during the firing of a big gun or when a projectile is fired against armor, it is found that steels that act alike at lower velocities may act differently and that brittle behavior may appear that would not be expected on the basis of static tests.

Again, there may be a size effect. A steel that shows toughness in all small scale tests may behave in brittle fashion in a very large section.

The appearance of brittleness is partly a function of the loading conditions and not wholly inherent in the steel. Steels do vary greatly in their behavior under the same loading conditions, but it would probably be possible to find conditions of loading under which almost any steel could be made to fail in a brittle fashion, though fortunately such conditions are very rare in service.

The first two of these variables in the loading conditions which bring out the inherent brittleness of the steel can be found from a static bending test as well as from the impact test; the third can be found only from an impact test, though the necessary striking velocity to induce brittleness is usually higher than can be obtained in most impact machines.

Most of the phenomena concerned are brought out even on a slow bend test, like the Humphrey,[75] but for speed and convenience the bars are generally broken by impact, and we can expect that sometimes the impact test will show more.

The differences in propensity toward brittle behavior are sought out and accentuated when the volume that can take part in the deformation is limited, i.e., when the energy of fracture must be absorbed in a very small volume. For example, if a tensile test is made on bars of the same steel with and without notches, the cross section at the base of the notch being the same as that of the unnotched bar, the tensile strength goes up and the elongation goes down in the notched bar. The energy absorbed drops rapidly, as Fig. 18 and the accompanying Table [76] show.

Obviously, a steel with high reduction of area in the regular tensile test tends to give a good account of itself under such conditions, but the correlation is not perfect.

However, no ordinary property of a steel other than perhaps the reduction of area seems to bear any relation to the notched-impact test, as Fig. 19 shows. Increase in tensile strength, Brinell hardness, yield strength, energy absorbed in the tensile test, tensile, impact, or endurance limit are not necessarily reflected in an improved notch-impact value. On the contrary, Fig. 19 seems to show that great

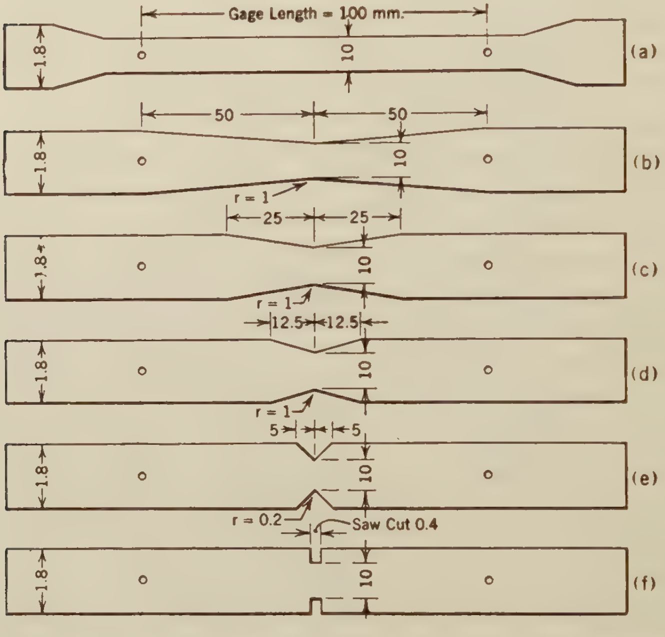


Fig. 18.—Notched cylindrical tensile test bars, in section. (Dimensions are in millimeters).

The notch effect in tensile tests, on bars shown above:

Bar	Tensile	Elong. %	R.A. %	Effective Elong.*	Energy absorbed, (kg.cm)
a	56,000	35.7	69.0	233	10,147
b	62,000	7.5	64.6	182.5	2,281
c	65,500	5.6	58.2	139	1,772
d	73,500	3.6	42.7	74.5	1,232
e	84,000	2.3	32.8	49	905
f	91,500	1.5	22.5	29.1	710

* $\frac{\text{R.A.}}{1-\text{R.A.}}$, Estimated effective % elongation. (Ludwik and Scheu, quoted by Hoyt [76])

strength or hardness can be obtained only at the sacrifice of toughness as measured by the notch-impact test.

Repeated Impact. Tests in which a round bar is grooved with a notch, which may be square, as in the usual Stanton test, or rounded, as in Knowlton's[77] tests, Fig. 19, have been used. The bar is supported at two points with the notch midway in the span, and a hammer dropped on the bar at the notch. The bar is turned over 180° between each blow, or, in some machines, through a smaller arc between blows. Depending on the intensity of the blow and the properties of the material, fracture will ensue at some moderately reproducible number of blows. The test is little used since it is generally thought that, if fracture occurs in relatively few blows, materials are rated about as by the single-blow, notched-bar test; and, if it occurs after many thousands of blows, they are rated much as by the notched fatigue test. No agreement is found as to standard conditions for making the test or as to interpretation of results obtained under such conditions that fracture occurs between say 1000 and 100,000 blows.

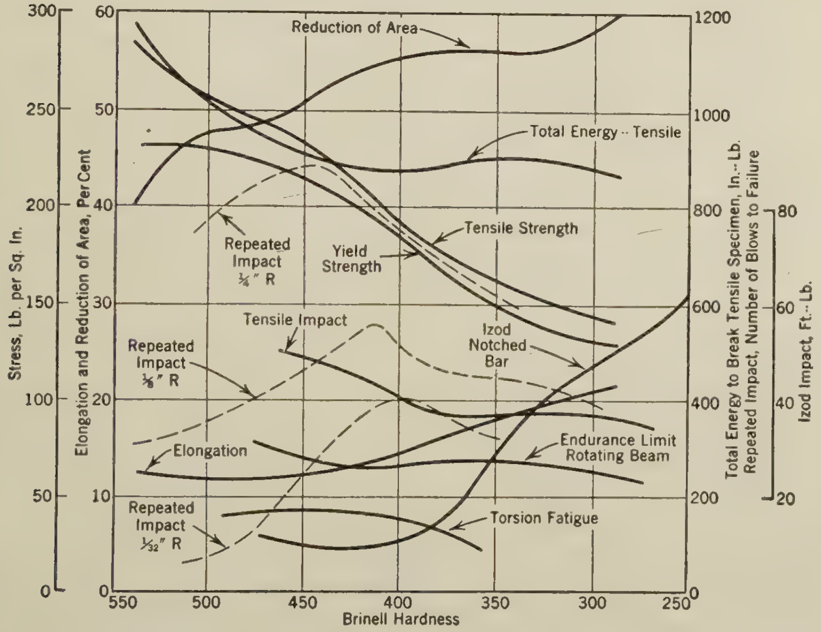


FIG. 19.—Mechanical properties of heat-treated S.A.E. 3140 steel, plotted in relation to hardness. (After Knowlton[77]).

The test was formerly used in the study of notch sensitivity of steels for automotive use and had some strong adherents. Latterly, it appears to have fallen into disrepute.

Machines and Specimens for Impact Testing. In order to bring out the propensity toward brittle behavior, standard tests have been devised in which a notched bar is fractured by bending, under a single impact. The two types of tests most used are the Izod and the Charpy. The energy absorbed in fracture is measured, as by attaching the striking hammer to a pendulum, which is allowed to swing from a known angle, to hit the specimen at the low point of its swing and to rise, in the remainder of its swing, to a smaller angle than it would have risen to had it not given up part of its energy in breaking the bar. From the angle to which it actually rises, the energy absorbed by the bar may be computed.

In the Izod test the bar is vertical and rigidly gripped at the lower end. The hammer hits above the notch and breaks the bar by cantilever action. In the Charpy test the bar rests at each end against rigid supports, the notch being in the middle and the hammer striking at the notch. The bar is thus broken in transverse bending. (See Fig. 21.)

The Charpy bar is square in cross section so that it may rest on a flat side against the supports. The notch for the Charpy is ordinarily made by drilling a small hole tangent to the center line of the bar and cutting down to that hole with a saw cut. Alternatively, the notch may be made as a V.

The Izod bar uses the V-notch and is usually square like the Charpy but may be round, with a flat milled on it to receive the striking hammer. The several bars are shown in Fig. 20.

The 10 × 10 mm. (0.394 × 0.394 in.) Charpy bar with a keyhole notch 5 mm. deep, leaving 10 mm. × 5 mm. (0.394 ×0.197 in.) back of the notch, has been adopted in most of the Continental countries.[78]

Results are expressed in meter-kilograms or foot-pounds of energy absorbed for a particular bar, the notch and the dimensions of the bar being stated. In the United States, results of either test are normally expressed in foot-pounds. In foreign literature, impact results are sometimes expressed in meter-kilograms per square centimeter; i.e., the energy absorbed is divided by the area back of the notch. Even when expressed in such units, results of tests made with different sizes of bars and different notches are not interconvertible. With the warning that conversion is *not* accurate, the following after Monypenny [79] is shown merely to give the order of magnitude of the differences and to indicate that the relationship is not linear.

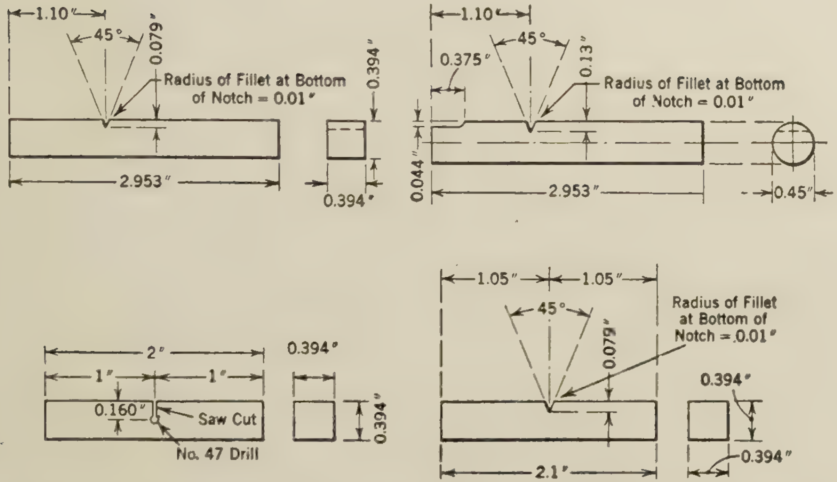


Fig. 20.—Some of the forms of notched bar impact specimens. Keyhole Charpy bar (preferred type), bottom, left. Alternative Charpy bar, bottom, right. Square Izod (preferred type), top left. Alternative round Izod, top right. With these various specimens in vogue, it is evident that a "Charpy" or "Izod" figure reported by different investigators does not necessarily relate to the same bar. Unless it is known that the same bar was used, data from different investigators may not be comparable.

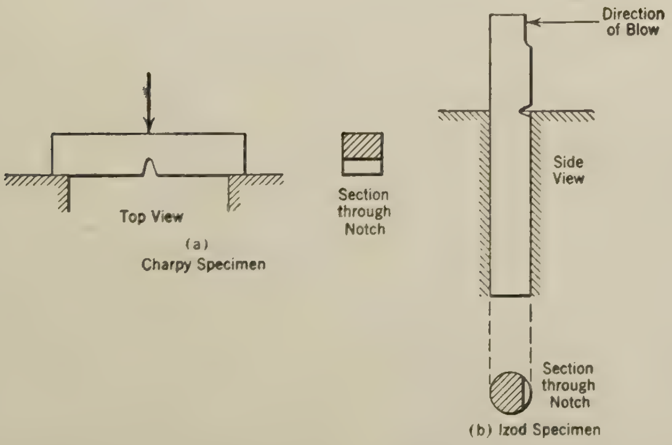


Fig. 21.—Method of loading single-blow, notched-bar impact specimens.

Charpy—meter-kilograms per square centimeter	2	4	8	12	16	24
Izod—foot-pounds	5	10	30	55	75	100

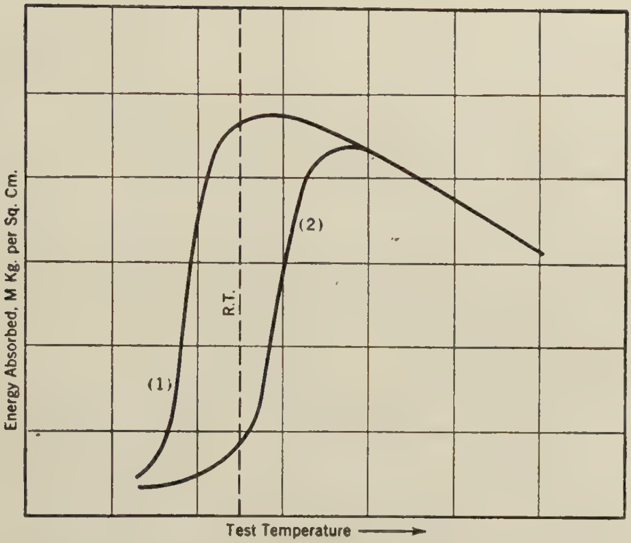
In general, the V-notch is more searching and tends to accentuate the propensity toward brittleness, though this is not always the case.

There is usually no material advantage in using one of the two types of test as compared with the other, since both usually tell similar stories. However, impact tests at high and low temperatures are becoming increasingly important, and the Charpy bar can be heated or cooled to the desired temperature, laid on the supports and broken without delay; whereas the Izod has to be clamped into the base, which takes time, and the base also has to be heated or cooled to the test temperature. Even when one is not primarily aiming at studying the properties at an elevated or lowered temperature, there is much value in testing over a range of temperatures rather than at room temperature alone; so it would be desirable to adopt the Charpy method whenever feasible.

The "A.S.M. Handbook"[80] states that the general opinion is that the keyhole notch is superior to the V-notch for general testing. On this basis the Charpy test, in which the commonly used bar has the keyhole notch, would be the logical one to select as standard. Unfortunately, in some industries, the available data are chiefly on Izod bars and specifications call for Izod tests, so Izod rather than Charpy data continue to be accumulated. This is the case with steel castings. The Izod bar is often made with several notches, on different faces, the bar being regripped after the fracture of each section. Less stock and less machining costs are required for duplicate determinations than when Charpy specimens are used.

Influence of Dimensions of Notched Bars. When either type of test and any one of the common types of notch are used, several peculiarities will be noted when the dimensions of the bar are altered. When the bar width is doubled, the energy absorbed may be much less than doubled and may even decrease. The difference varies with different steels, and one method of evaluation is to find the ratio between narrow and wide bars. This behavior, diagrammatically shown in Fig. 22 from Hoyt,[76] shows clearly that a single impact figure made on a "standard" specimen is not necessarily an adequate criterion for evaluating differences in steels.

For instance, two steels that really differ may behave in a ductile fashion in the test and give the same energy absorbed. This is all the



(After Hoyt[76])

FIG. 22.—Schematic representation of effect of narrow (1) vs. wide (2) bar, or of fine grain (1) vs. coarse grain (2) in steel, upon impact results over a range of temperature. With many steels the rapid drop in impact resistance occurs near room temperature. At the higher temperatures the fracture is fibrous, at temperatures below the rapid drop it is crystalline.

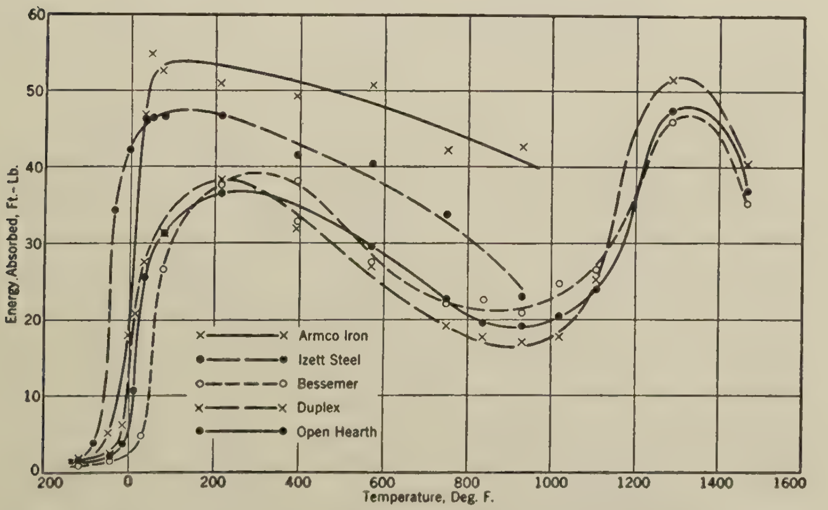


FIG. 23.—Charpy impact data (keyhole notch) for low-carbon structural steels. "Izett" is a fine-grained, aluminum-treated steel. (Epstein[81]).

test does tell. It does not tell us whether any difference exists in inherent brittleness of the steels, and to find this we would have to increase the severity of the loading conditions until one was ductile and the other not.

Effect of Temperature on Impact Resistance. Using the same bar, it is found that a change in testing temperature may radically alter the results, as is shown in Fig. 22 as well as Fig. 23.[81] Some commercial steels show this very sharp drop in impact resistance around room temperature, and when this occurs duplicate specimens may give widely different values of energy absorbed. Until it was realized that the impact-temperature curve showed this sudden drop and that erratic test results must follow when the test temperature coincides with it, impact results were considered to be so non-reproducible as to be worthless. As a matter of fact, in regions of the impact curve where no sudden change is occurring, impact tests are as reproducible as tensile tests.

The S-shaped curves, such as those shown by Epstein, Fig. 23, are characteristic of the impact properties of practically all steels. At some range of high temperature, impact resistance will be at a high level and, at some very low temperature, at a very low level. These high and low levels may be higher or lower with different steels, and the curve joining them may rise steeply or gradually; but an S-type of curve, rather than a straight line, is normal. Although steels quite high in Ni and a few other types, including several austenitic steels, are notable exceptions, most ferritic steels become very notch brittle at liquid-air temperature and almost equally so, though they may be at quite different levels, at room temperature. Exceptions will be noted in Chap. 19, Vol. II.

Conversion of Charpy and Izod. Results of Charpy and Izod tests cannot be accurately converted one into the other over a range of materials. Under very restricted conditions, conversion is sometimes attempted. Foster [82] gives some conversion factors and Aldridge and Shepherd [83] worked out empirical conversion factors for low-temperature impact testing on Ni steels. They also compared results on specimens of varying thickness so as to be able to test plate too thin to provide a standard test specimen. Epstein, Nead, and Halley [50] made low-temperature impact tests on ⅛-in. and $\frac{1}{16}$-in. sheets. Comparisons may of course be made within any group of materials by use of non-standard specimens, but conversion of the results for comparison with results on standard specimens is attended with uncertainty.

Impact as a Mirror of Grain Size. A high level of impact resistance at room temperature and a postponement of the sharp loss in impact resistance to decidedly lower temperatures are generally brought about by fine grain size.

No mechanical test is so sensitive to grain size as the impact test. Since properties more troublesome to test for than impact, are likewise connected with grain size, the impact test has ability as a convenient mirror of the grain size, quite aside from what it tells about brittleness.

Indeed, it is as a mirror of other difficultly determinable properties or propensities that impact results are perhaps more useful. Temper brittleness, for example, is revealed only by impact.

Relation of Impact to Service. How much value the impact test has in evaluating materials for service, from the point of view of selecting one material instead of another, is a moot point. It has been said that landing gears for airplanes stand up in service better when made from materials that have high impact resistance, and rail failures in cold weather are thought to be related to the very low impact resistance shown by much rail steel in the range of —[illegible]°. Nevertheless, when proper fillets and general avoidance of sharp notches are used in design, materials of low impact resistance may serve quite as well as very tough ones.

The service toughness of different classes of materials is not necessarily correctly compared by their impact values. Most classes of steel would be considered dangerously brittle with an impact resistance of less than 10 ft.-lb. Izod, yet spring steel in spring temper and ordinary rail steel will seldom show as much as that. Malleable cast iron is said [85] to be so ductile that it seldom or never fails in compression or transverse loading, since it flows indefinitely, or deforms so much that the part is useless, before it fractures; yet malleable will nevertheless normally run only about 7 to 12 ft.-lb. Izod or Charpy, and in appraising "galvanizing embrittlement" of malleable a figure of 5 Izod or 7 Charpy is considered to show a tough condition, not a brittle one. Malleable is used for shock-resisting service in railway uses, in road machinery, agricultural implements, and the like. Even when the specifications for elongation in the tensile test called for only 10%, as they did in 1924, instead of the 18% now required in the highest grade malleable, the material was considered tough and suitable for shock-resisting uses.

Rail steel subjected to different heat treatments has been tested by impact on un-notched bars in order to secure a higher level of values and make results more discriminatory.[86] Un-notched bars are used, for the same reason, in testing alloys for zinc base die castings, but in general such tests are not thought to indicate behavior under shock as definitely as do tests on notched bars. Tensile-impact tests,[87] in which an un-notched tensile bar is rapidly broken by shock loading, show differences among steels, but data so obtained are too scanty to allow evaluating the method.

In the study of tool steel, and of toughness on very hard steels (say Rockwell "C" 60 and up, i.e., where the Charpy and Izod figures are so low that the tests are not discriminatory), a torsion impact test on an un-notched bar has been devised, but its meaning and interpretation are not yet certain.[88]

Impact values in themselves are of no direct value in design. A material of 50 ft.-lb. impact is not ten times as good as one of 5 ft.-lb. The two may be exactly as good, or the low one may be of no use at all.

When impact values upon a material, so heat treated that it gives an average of 40 ft.-lb., rise to 45 or fall to 35, it means little. But when the values fall to 10 or 15, it means that something has occurred that is not according to schedule and we had better look into it. Similarly, if a material that has been in long-time, high-temperature service is removed, and tested at room temperature, and shows no change in tensile properties, but does show its impact to be halved or quartered, it means that the material is not wholly stable and that we had better find out what is going on. Thus, the impact test finds its chief use in telling the metallurgist something, at least, about variations that he might not otherwise suspect and serves as a signpost to indicate the necessity for further study of these variations.

The relation of impact resistance to service has been much discussed. Thum [89] has summarized some of this discussion, and a British symposium was devoted to it.[90, 91, 92] The American Society for Testing Materials held a similar symposium in 1938.

It is interesting to note that in both symposia it came out clearly that the absolute value of impact resistance in lb./in.^{2} is of less importance than the type of fracture. A fibrous fracture indicates a tough steel, a crystalline fracture, a brittle one. The conditions of service and the type of steel both control the transition from tough to brittle.

The paper by Riegel and Vaughn [132] in the A.S.T.M. symposium shows that impact testing, combined with engineering experience in its interpertation, has allowed differentiation between satisfactory and unsatisfactory steels for heavily stressed gears in tractors, when chemical analysis and static properties failed to distinguish between them.

Qualitatively, there is virtue in toughness, and the impact test certainly tells something about toughness. The test may need to be made with a variety of notches, on specimens of various sizes, at different temperatures and different rates of loading, in order to make it tell a more complete story. The exact figure required to pass a steel for a given service will continue to be a matter of debate, but increasingly the metallurgist is being required to produce material to meet an impact specification based on the impact test results that have been shown by material in satisfactory service. Control of heat treatment so as to produce specified impact resistance is therefore increasingly necessary.

Bend Tests. The ability of a steel to be formed by cold bending is an important property to the fabricator. The bend test measures cold formability quite directly. Expensive equipment is not required in the test as usually carried out; a flat test specimen is, for very thin material, bent over flat upon itself, or, for thicker stock, over a pin whose diameter is specified. The diameter is usually stated as some multiple of the thickness of the specimen, the multiple increasing as the severity of the test is decreased. The bent specimen is examined for cracks.

The test may be refined by determining the local elongation on the outside of the bend, as suggested by Kinzel [93] and used in the study of welds. For thin stock, repeated reversed bending may be used as a criterion of formability and toughness.[50, 94–96] Wrap tests on wires are similar direct tests for the ability of a wire to stand the type of deformation it may receive in service.

Magnetic Testing. Certain magnetic tests are employed in the evaluation of material for use in permanent magnets, transformers, and other electrical equipment, too specialized a matter to discuss here. From the point of view of heat-treatment problems, it is sometimes of interest to know the amount of austenite retained after quenching; and this can be shown by magnetic methods.

A valuable method for making apparent such defects as hardening cracks, grinding cracks, cracks developing in service, and other flaws of analogous types is to magnetize the object and note the way fine cast-iron dust suspended in kerosene is attracted to the specimen. Cracks not otherwise visible are thus made evident. This general method has been specially developed by de Forest under the name of "Magnaflux" inspection [97, 98] and is finding considerable acceptance.

Radiography. Radiographic inspection by X-ray or by gamma rays from radium is widely useful in detecting gross flaws in castings and welded structures, but since this does not throw light on variables in heat treatment, it will not be discussed here.

Damping. The ability of material to dissipate energy by converting it into heat within the material is known as its damping capacity. A steel tuning fork does not dissipate much energy and keeps on vibrating for a long time with only slow diminution in amplitude, whereas a similar fork of cast iron ceases vibrating very soon. Damping capacity varies with structure and heat treatment. It has been studied in relation to fatigue without noteworthy results but may perhaps be a useful tool in the study of the effects of reheating after cold work and of structural changes resulting from heat treatment.[99-101]

Cupping Tests. The ability of steel sheet to take deformation in cold pressing and deep drawing is affected by the structure and hence by the heat treatment of the sheet. Various cupping tests, such as the Erichsen and Olsen, are used in the effort to evaluate deep-drawing properties and have some value as rough indicators. Satisfactory laboratory evaluation of the properties wanted in actual forming is a difficult matter, on which little agreement has yet been reached.[102]

Weldability. Welding is so important a method of fabrication that the suitability of steels for welding bulks large in their selection. Welding is very seldom used on heat-treated steels, since the welding heat alters the structure that has been secured by heat treating. Heat treatment would have to be applied after the structure has been assembled, and most welded structures are too large to be conveniently heat treated. Even a mere stress-relief anneal on large assemblies involves expense and difficulty.

The relation of heat treating to welding ordinarily lies in the selection of steels for welding that are *not* too readily heat treatable, i.e., those that have little propensity toward air hardening on cooling from the welding heat, so that the welds and the stock adjacent thereto are not embrittled. The American Welding Society is publishing a handbook [103] which should be consulted on this topic.

Machinability. Since practically every piece of steel that is heat treated is machined before or after treatment, or both, the ability of the steel to be readily and cheaply machined is often of paramount importance; and steel is often chosen or rejected because it does or does not fit the requirements as to machinability. A quantitative measure of machinability would be a boon, but seldom indeed can anything better than a rough qualitative evaluation of this property be obtained. In different types of machining, such as turning, drilling, reaming, milling, etc., different steels and different structures stand in different orders. The cutting tool, in its own composition and structure, in its geometric form, and in the speed and feed at which it meets the work, affects the machinability of the work. Even in grinding, the particular abrasive used, the size and shape of the particles, and the nature of the bond have to be chosen for the particular job. Moreover, the chief criterion of machinability may be the speed with which the steel can be machined, the life of the cutting tool, or the quality of finish; and all these may enter into the overall cost, which is the ultimate criterion.

Wide differences in behavior on machining can be detected and described, but small differences, such as those between different grades of "free-machining" steels, are very difficult to evaluate outside the machine shop itself, and even then different shops may rate a series of steels in quite different order. (Compare Vol. II, Chap. 2.) Armour[104] has given a good discussion of machinability.

In spite of the difficulty of exact appraisal of machinability, gross differences may be appraised by simple tests. "Sawability," determined by noting the time to saw through a bar of steel, chosen as the standard for comparison, and the time to saw through the same sized bar of the steel being tested under the same pressure on a power hack saw, and "drillability," similarly determined by time to drill a given depth into a comparison standard and the steel being tested, using a power drill press and the same pressure, are useful determinations. Much care has to be taken to ensure that the saws or drills used have the same cutting power in a series of tests, so taking a cut on the standard steel before making the test and again after the test, to make sure that the tool was normal at the start and only normally dulled during the test, is an important precaution. Uniformity of the comparison steel chosen is obviously necessary.

Wear Resistance. Resistance to abrasion by non-metallic materials and to attrition by one metallic part moving upon another is required in many services, but wear resistance, like corrosion resistance, is a function not only of the steel itself but equally of the conditions of service.

There is no inherent and measurable wear resistance of steel as an innate property, and correct promises of performance in service can only be had in the laboratory by devising a specific wear test reflecting the particular service condition to be met. An A.S.T.M. symposium[105] has dealt with wear and wear testing.

As a first approximation, wear resistance tends to rise with hardness, as long as this does not involve extreme brittleness and chipping or spalling of the surface. The presence of finely divided and uniformly distributed particles of free carbides in a softer matrix adds resistance to most types of wear, especially when the carbides are of a particularly hard nature, as in Cr steels. A sufficient amount of C to provide for the presence of such particles is therefore found in most tool and die steels.

When the surface pressure is great enough to produce a work-hardened skin, such as is characteristic of austenitic steels, particularly of the high Mn steels (Vol. II, Chap. 20), good wear resistance is produced in these steels but if the service is not of this type they perform poorly.

Work Sensitivity. The rate at which a steel work hardens under cold deformation has a bearing on its machinability, its resistance to some types of wear, and its suitability for cold forming. The shape of the stress-strain curve beyond the yield stress tells something about this property, and special, but as yet unstandardized, tests have been devised to evaluate it.[106–107]

Castability. A cast steel that will readily flow into thin sections has advantages over one that flows with difficulty. Between two steels that show identical properties in the solidified casting, the one that is the more fluid would be chosen because of its better behavior in the foundry. "Castability" may be determined by casting a thin strip or spiral at a definite temperature and under definitely controlled conditions and noting how far the metal runs along the strip or spiral.[108]

Elusiveness of Unmeasurable Properties. If all the properties desired in steel could be quantitatively measured and if the user could state definitely just what degree of each property his service demanded, the evaluation of steel and its heat treatment to give it the desired combination of qualities would be easy. A formula could be set up in which each measured property was weighted as to its importance and a final figure of merit arrived at. Indeed, many "merit indices" have been proposed, to little avail.

In our present stage of ignorance on the testing of steel, the procedure is for the user to measure what properties he can, to observe how steels varying in those properties serve his needs, and to specify as well as he can what minimum properties he thinks he wants, in order to have assurance that the performance of another piece of steel will duplicate satisfactory past performance. In this he is likely to demand, just for security, properties that are not truly essential to his service, and thus to raise the cost through the expense for selection, treatment, and inspection for unnecessary qualities. However, until more quantitative measurement of the necessary, but elusive, qualities is feasible, this is the only avenue open to him.

Statistical Methods. The consistency with which a given steel, put through a given heat treatment, will produce the same results is a matter with which the engineer is very deeply concerned. If a low value may be expected to occur now and then, inspection testing must be done on enough samples to minimize the risk that such a piece has slipped through. Lacking the assurance that low values will not be met, the engineer has to design on the basis of the lowest value he thinks will be met, even if but 1 out of 100 shows that low value, so he cannot make use of the better properties of the other 99. This hampers the aircraft designer in particular, where material of great uniformity is sought so that the factors of safety may be held at a minimum.

It has been well said that "consistent mediocrity is to be preferred to intermittent perfection." Because the designer has to base his design on minimum values, the Society of Automotive Engineers' curves [128] of physical properties of steels are drawn for the *minimum* tensile, yield, elongation, and reduction of area, though, in a given specimen, when the ductility is at the low point, the strength is usually above the minimum.

There are ways of evaluating the probability of getting values lower or higher than the average if sufficient specimens are tested, and of telling what is a "sufficient" number of specimens. We may plot the points, draw in the usual curve to denote the average, and also draw a band enclosing all the high and all the low points. The width of this "scatter band," and whether the individual points are scattered widely or tend to concentrate closely above the average curve, tell us something. The data may be plotted in another way, i.e., according to the frequency of occurrence. A concrete example of these types of plotting is shown, in data here reproduced by courtesy of the S.A.E., in Figs. 24-27 from a report by Janitsky [109, 110] for the tensile and hardness properties, on 4 lots of S.A.E. 6130 CrV steel. Each of the 4 steels was made by a different steel company, to accord with the regular specifications. Three 1-in. rounds of each steel, each long enough for 4 tensile tests, were sent to 30 laboratories, with definite instructions for quenching and tempering at 3 temperatures. The 120 final collected values for each property at each temperature, then reflect all the variations in composition of the steels, variations in heat-treating temperatures and in testing technique, calibration of testing outfits, and similar variables. A surprisingly wide variation was found, and the results of some investigators were so obviously erroneous that they were excluded from consideration in the plots shown. Possibly their pyrometers were off calibration.

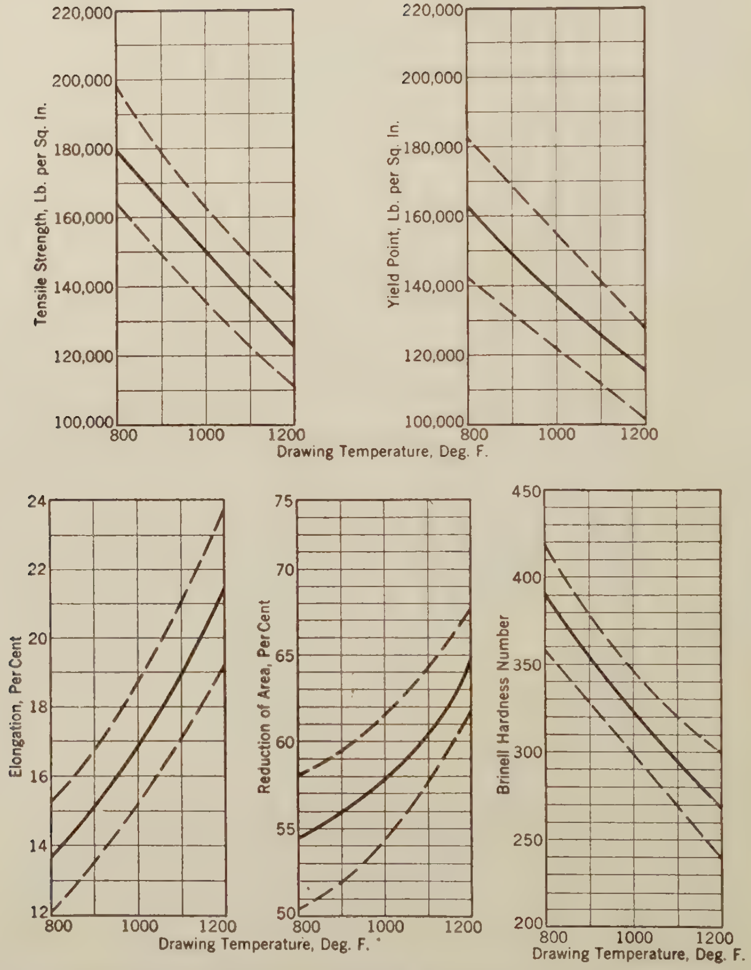


FIG. 24.—Summary of mechanical properties of S.A.E. Cr V steel 6130. Maximum and minimum values are represented by dashed lines and averages by solid lines. This is the "scatter-band" type of plot. (Janitsky[109, 110] S.A.E.).

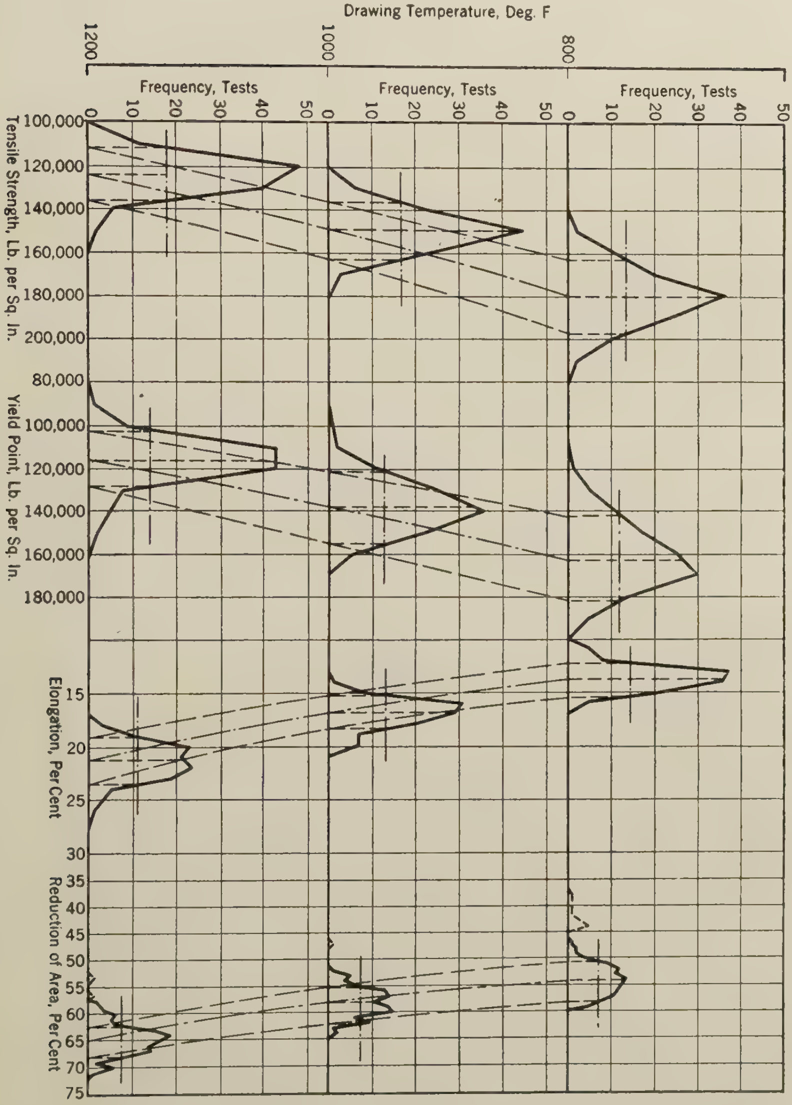


FIG. 25.—Mechanical properties of S.A.E. Cr V steel 6130 drawn at 800° (right), 1000° (center), and 1200° (left), plotted as frequency curves. (Janitsky [109, 110] S. A. E.).

According to the advocates of V[111] its use tends towards uniformity in properties from heat to heat, and this is reasonable in view of present knowledge of its effect upon grain size, as its use would tend to avoid variations in grain-coarsening tendencies that might otherwise occur. On this basis, had other than a V steel been chosen for these studies, and other conditions been the same the results might have been even more erratic.

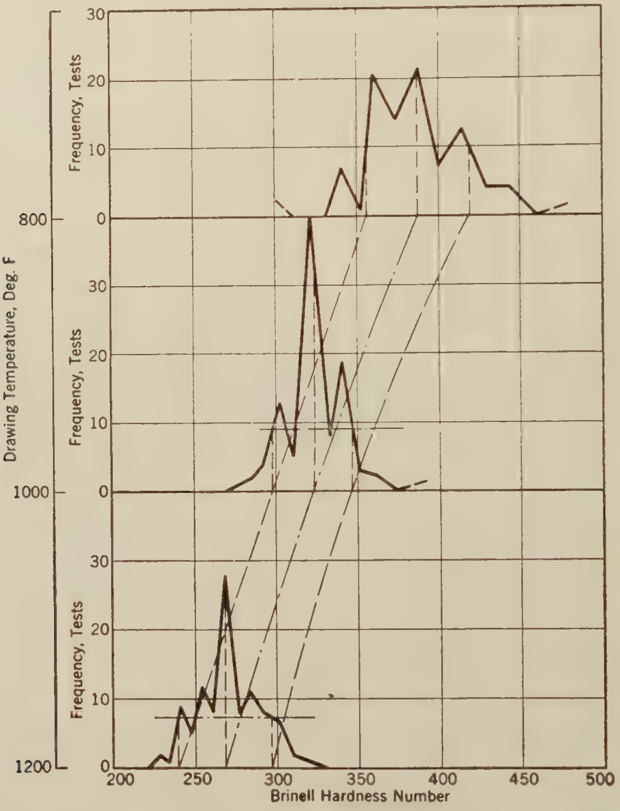
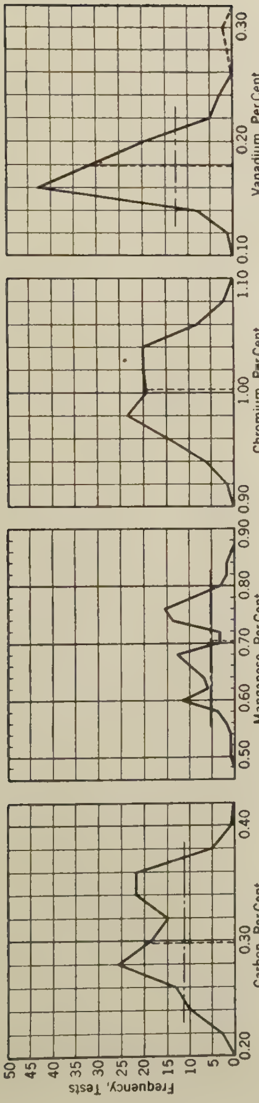
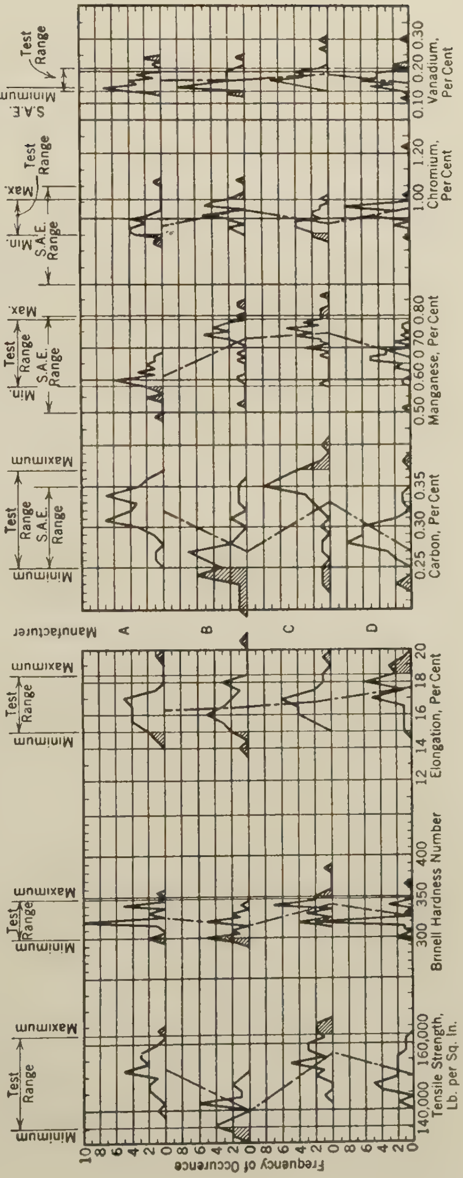


Fig. 26.—Frequency curves of hardness tests of S.A.E. Cr V steel 6130 drawn at 800° 1000° and 1200°. (Janitsky [109, 110], S.A.E.)

When the data were similarly studied on the basis of the 4 individual steels, as shown in Fig. 27*b*, rather than lumped as S. A. E. 6130, it became evident that the differences in carbon and chromium contents, though within the specification limits, were exerting definite influence. Similar data[110] for the Ni Cr steel, S. A. E. 3130, are shown in Figs. 28 and 29.



(*a*)



(*b*)

FIG. 27(*a*).—Frequency curves for chemical analyses of the 4 steels whose properties have been lumped in Figs. 24–26.

FIG. 27(*b*).—Frequency curves for analyses and for corresponding mechanical properties, of the 4 steels, separately. These 4 Cr V S.A.E. 6130 steels were water quenched from 1600°. Fig. 27(*b*) shows data only for the 1000° draw. (Janitsky,[109, 110] S.A.E.)

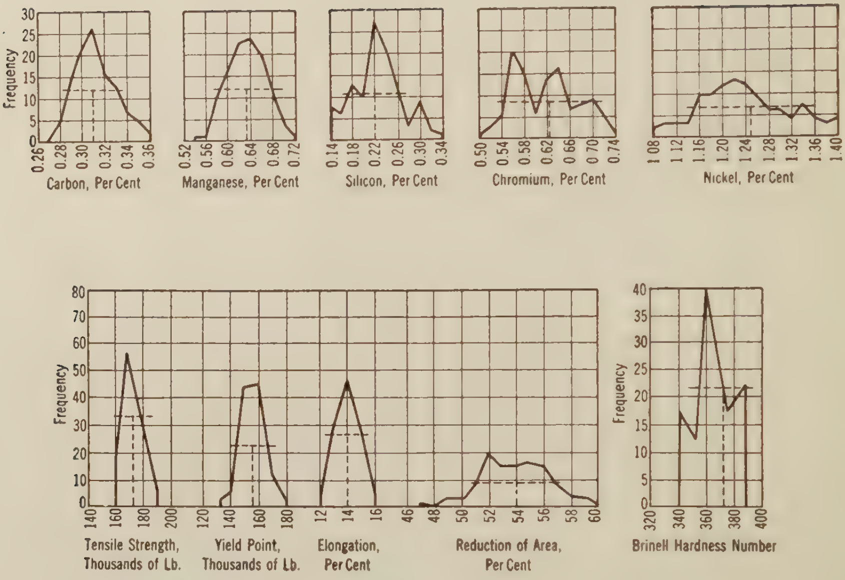


FIG. 28.—Frequency charts for Ni Cr S.A.E. 3130 steel. Chemical composition at top: mechanical properties, water quenched from 1525°, drawn at 800°, at bottom. (Janitsky [110], S.A.E.).

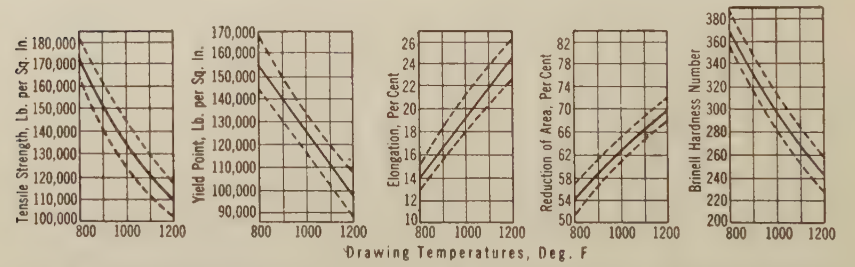
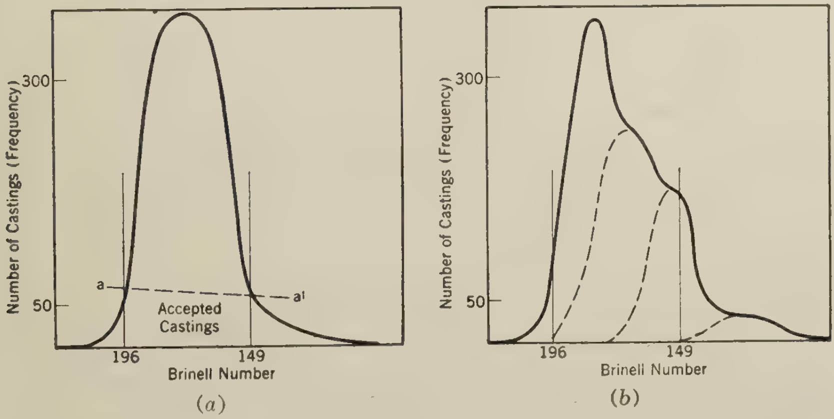


FIG. 29.—Scatter-band plots for the Ni Cr S.A.E. 3130 steel at draw temperatures from 800° to 1200°. (Janitsky [110], S.A.E.).

Probability plotting, so that the steepness of the hump in the curve is indicated, as well as whether it has only one peak or more than one, often points the way to the study of variables that had not previously been considered important and indicates what variables should be separately studied with everything else rigorously held constant. Saklatwalla and Chandler[112] cite the application of probability plotting to a foundry engaged in casting flywheels, where a certain Brinell specification had to be met. Their Brinell readings for one month are plotted in a probability curve in Fig. 30*a*. This shows that their composition and practice are nearly right for meeting the specifications. The only fault would be that the iron had somewhat too wide limits of hardness. However, conditions are not bad; the percentage of rejects is about 8. Figure 30*b* shows the results for the following month, plotted in the same way. Here it is seen that the probability curve no longer has a regular shape. This indicates that the probability curve is a summation of several, different, curves indicated by the dotted lines. These results show that there are certain influences, apparently three, which are tending toward the production of softer iron. These influences reflect some changes in foundry procedure, and a knowledge of what has been happening in the foundry would enable these changes to be corrected. A mere average of the hardness results would not tell anything, since the average hardness of the results is the same within two points in Fig. 30*a* as in Fig. 30*b*.



(*a*) (*b*)

FIG. 30.—Probability curves of Brinnell hardness of cast iron. (Saklatwalla[8] and Chandler[112].)

Study of the frequency curves often brings out the fact that a set of observations from which we are hoping to arrive at the normal behavior and the range in behavior may be too limited in number to give true indications of either. Likewise we may be sampling or testing too small a proportion of our output to keep from blinding ourselves to the possibility of defective material being shipped and to the risks we are taking of having that defective material show up in service, come back, and injure our reputation. Statistical methods properly applied and used from a common sense rather than too highly mathematical a viewpoint, can clarify these problems.

Statistical methods are valuable for determining the suitability of an alloy for meeting certain specifications. From probability curves we can find not only what proportion of our output is meeting specifications but also whether its composition and treatment have been right to make the average come in the middle of the specification range. Also, a lop-sided probability curve will show that there are certain influences present whose correction is necessary in order to obtain more consistent results.

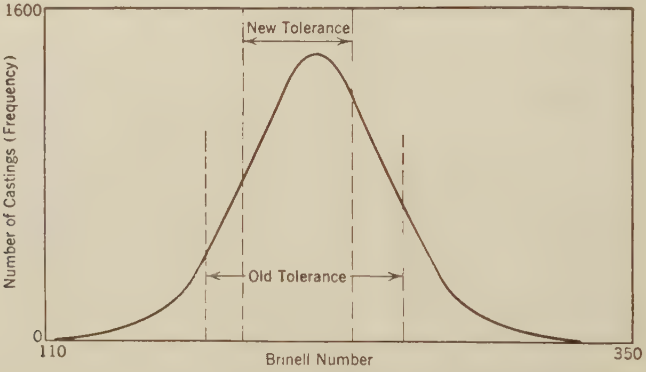


Fig. 31.—Example of use of probability curve in determining the effect of a change in specifications on the proportion of the product that will meet the new specifications. (Daeves[116]).

If only non-destructive tests, such as hardness, are involved in meeting specifications, we can, of course, test each piece in order to be sure that none of the output will fail to meet the requirements. When destructive tests, such as the tensile test, must be performed on the object itself, probability curves indicate what percentage of the output will meet specifications and how technique should be varied the better to meet specifications.

Where new, more rigid specifications have just been required and a cost must be figured, Daeves in discussion of a paper by Saklatwalla and Chandler[112] shows how this can be done by probability studies. This method is illustrated in Fig. 31. The old tolerance and the cost of making it are known. From the figure we can determine the percentage of parts falling within the old specification limit, and by finding, from the curve, the percentage of parts falling within the new specification, the new cost can readily be figured.

Such methods have been applied by Pulsifer[113] to the study of scatter of properties of heat-treated bolts. Chancellor[114] and Waterman[115] discussed the topic in relation to variables in steel making, whereas Daeves[116] uses many metallurgical examples in setting forth the general principles. An article by Hoar[117] is concerned with metallurgical applications. Related to sampling and testing are the "Manual on Presentation of Data"[118] and the discussion by Dodge.[119] The books of Shewhart,[120] Pearson,[121] Levy and Roth,[122] and Snedecor[123] are of interest to the advanced student of the subject.

Even a slight acquaintance with the probability point of view, which can be gained without going into difficult mathematics, is so illuminating to those who make tests or draw conclusions from test data that it may well be classed as an essential part of the information of the modern testing engineer as is an understanding of the fact that a tensile machine that is out of calibration will not give reliable results.

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CHAPTER 2

THE STRUCTURE OF STEEL

Steel. Steel is an alloy, the principal and most essential chemical constituents of which are Fe and C. With these there usually are certain other elements such as Mn and Si, added for the definite purpose of deoxidation, small amounts of P and S, as well as generally a little Cu and Ni, and sometimes Cr and other elements. Of the elements which go to make up ordinary steel, the Mn, Si, P, and S usually total from ½ to about 1% ; the C will vary from a few hundredths to about 1.7%.

Steel is not a uniform substance like pure copper or gold, but is more like granite, in that it is made up of a number of individual grains corresponding to the minerals—quartz, mica, and feldspar—of the granite. Thus, in steel that has cooled slowly from a high temperature, we have "ferrite," "cementite," and "pearlite." And just as the relative amounts of the quartz, mica, and feldspar may vary in the rocks of the granite class, so will the relative proportions of ferrite, cementite, and pearlite vary in different steels according to the chemical composition of the steel.

Ferrite. Ferrite is the name given to Fe in the crystal form that exists at room temperature in slowly cooled C steel. It comes from the Latin word *"ferrum,"* meaning iron.

Cementite. The C and Fe are the essential and controlling elements in the steel—and this particularly is true of the C. In steels that have cooled slowly from a high temperature, the C is nearly completely combined with a definite amount of Fe to form a "carbide of iron," corresponding to the chemical symbol Fe_3C. This compound consists of 6.6% C and 93.4% Fe, and is called "cementite." (The term is derived from the high C steel made by the old "cementation" process). The remainder of the steel is practically C-free ferrite.

Alloying elements may be added for definite purposes. These are present in the ferrite, dissolved in it, or in the cementite, as complex carbides, or in both. The alloys modify the behavior of ferrite and cementite, but not to such an extent as to prevent the consideration of all steels from the standpoint of the plain C steel.

Pearlite. During the process of cooling at a slow rate from a red heat, cementite forms a mechanical mixture with ferrite. The mixture has a strong tendency to contain a definite amount of ferrite, so that the resultant normally contains approximately 0.80–0.85% C. The figure is not entirely certain. For ordinary purposes it may be rounded off to 0.85%. It is generally considered to lie between 0.78 and 0.88%. If the Mn is on the high side the eutectoid may contain less than 0.80%.

This constituent is called "pearlite" because it shows a pearly luster under the microscope, and usually consists of interstratified layers or bands of ferrite and cementite as is shown in Fig 1. In this micrograph, the cementite is dark, the ferrite white. Pearlite is regarded as a separate structure in steel, as it forms distinct "grains" when present in any appreciable quantity; it tends toward containing this fairly definite percentage of, say, 0.85% C, and—as will be explained later—is always born at a definite range of temperatures.

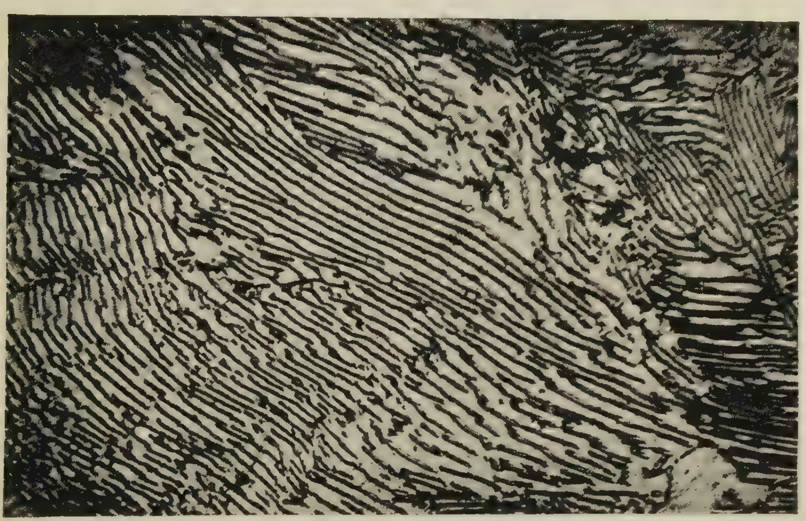


FIG. 1.—Pearlite in hypo-eutectoid steel. × 650. (Bullens.)

Eutectoid Steels. From this it will be seen that a steel containing 0.85% C will consist entirely of pearlite. Such steels are known as "eutectoid" steels, and that ratio of C as the eutectoid ratio.

Hypoeutectoid Steels. Steels containing less than this eutectoid ratio of C will consist of a definite amount of pearlite, varying according to the C content of the steel proper, and the remainder is "free" or "excess" ferrite. These steels are called "hypoeutectoid" steels, as they contain less than 0.85% C ("hypo" means "under").

Hypereutectoid Steels. Similarly, if the C content exceeds 0.85% C, there will not be sufficient ferrite to interstratify with all the cementite, so that these steels will consist of pearlite plus free cementite. Such steels are called "hypereutectoid" steels ("hyper" means "over").

Expressing this in a different way, very low C steels are made up of ferrite with a little pearlite. With increase in the C content of the steel, the amount of pearlite likewise will increase, with a corresponding diminution in the amount of free ferrite, until at 0.85% C the steel will be wholly pearlitic. Beyond this point the amount of pearlite will decrease, with a corresponding increase in the amount of free cementite.

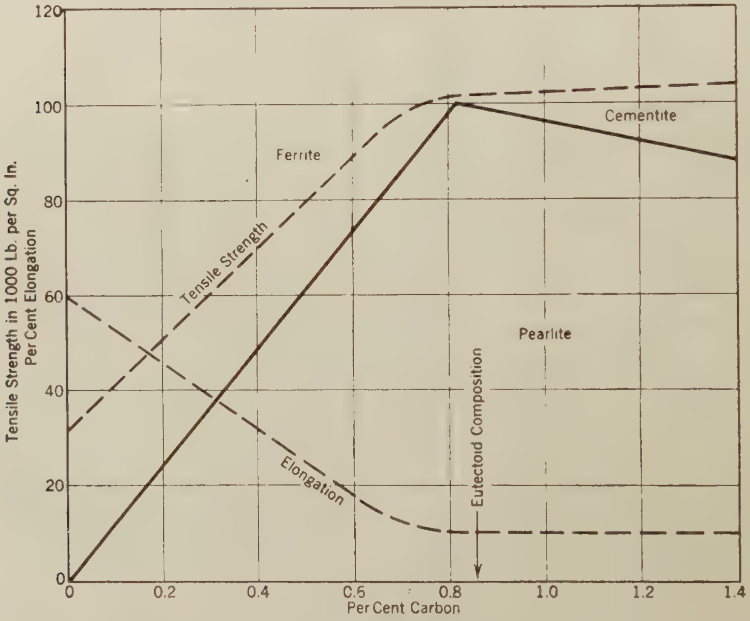
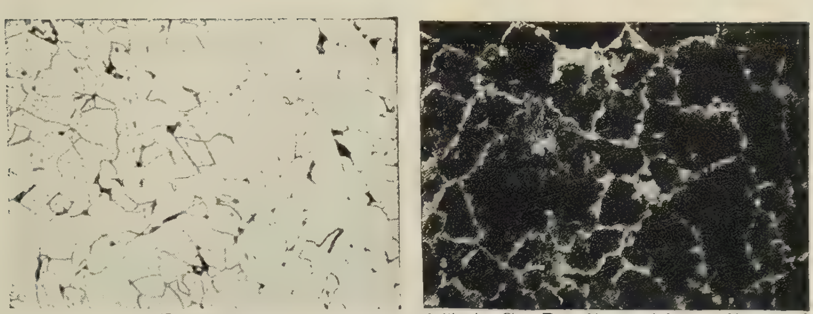


FIG. 2.—Ferrite-pearlite-cementite diagram (solid lines), compared with tensile strength and elongation of annealed steels.

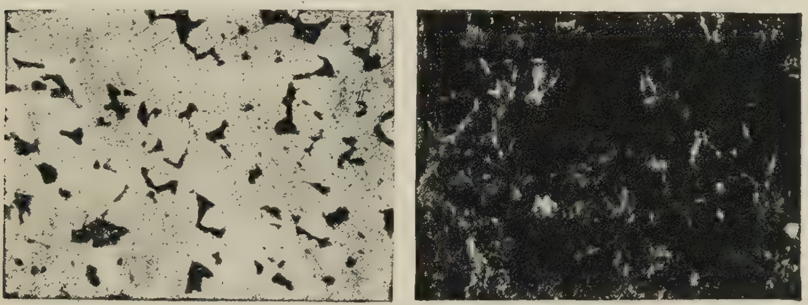
Structure of Slowly Cooled Steels. In slowly cooled steels, therefore, we may tell with fair accuracy the approximate structural composition of steel. And, *vice versa*, knowing the relative proportions of pearlite and ferrite or cementite, as determined microscopically, we may determine the approximate C content of the steel. This is represented graphically in Fig. 2.

These facts also are illustrated by Fig. 3 representing the structure



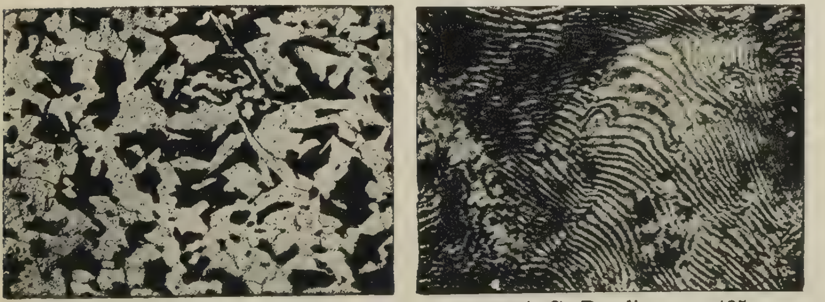
(Battelle)
0.09% C. Nearly pure ferrite.

0.57% C. Pearlite with outlines of ferrite.



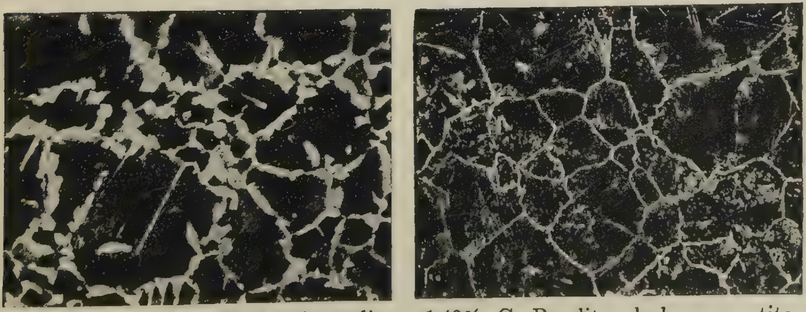
0.18% C.

0.71% C. Pearlite with traces of ferrite.



0.32% C.

0.83% C. Pearlite. × 485.



0.49% C. Ferrite is white and pearlite dark in this column.

1.46% C. Pearlite, dark; cementite, white.

× 75, Ordnance Dept. unless noted.

FIG. 3.—Structures of annealed steels of different C content.

of slowly cooled steels ranging from 0.09 to 1.46% C. The pearlite (dark; its duplex structure is not resolved at X75 *) gradually increases in amount, while the ferrite (light) proportionally diminishes as the C increases. At 0.83%, about the eutectoid composition, the ferrite (dark) and the cementite (light) are interstratified with each other, there being practically no "free" ferrite such as characterized the lower-C steels. The structure of the eutectoid steel is shown at X485 in order to resolve the structure of the pearlite. The structure of the 1.4% C steel shows the free cementite (white) which occurs as a network between the grains of pearlite (dark). In Fig. 4 we have nearly this whole range represented by means of case carburizing a steel of 0.17% C.

Physical Properties Dependent upon Constituents. Upon the relative proportions of these constituents will depend the physical properties of the slowly cooled steel, neglecting for the time being their relative arrangement. Each of these components—ferrite, pearlite, and cementite—has certain physical characteristics with which we must be familiar in order to gain some idea of the properties of such steels.

Ferrite. Ferrite is soft, ductile, and relatively weak. It has a tensile strength of approximately 40,000 to 50,000, with an elongation of about 40%. Ferrite has a hardness of only about 90 Brinell, and by itself has no hardening power applicable to industrial purposes. It is magnetic and has a high electric conductivity. Its appearance under the microscope has been shown in the ground mass of the 0.09% C steel of Fig. 3, that is, as polyhedral crystals in the low-C steels. That the material is crystalline is evident from the slowly cooled Fe, deeply etched, illustrated in Fig. 5.

Pearlite. As previously mentioned, the common occurrence of pearlite in slowly cooled steels is in the stratified lamellar formation, composed of alternate plates of ferrite and cementite. As will be shown later, under different rates of cooling, pearlite may exist in other formations dependent upon the relative arrangement of the ferrite and cementite of which it is composed. Normal pearlite, that is, interstratified bands of ferrite and cementite such as shown in Fig. 1, has a tensile strength of approximately 125,000 to 150,000, with an elongation of about 10%. Its hardness is 250 to 300 Brinell, depending on the fineness of its structure. The finer it is, the stronger.

Cementite. The properties of cementite are very little known, with the exception of its great hardness, about 650 Brinell, and its brittleness, which is extreme. Its tensile strength should be around

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* X75 or 75X is a notation used to show the number of diameters magnification; in this case 75 diameters.

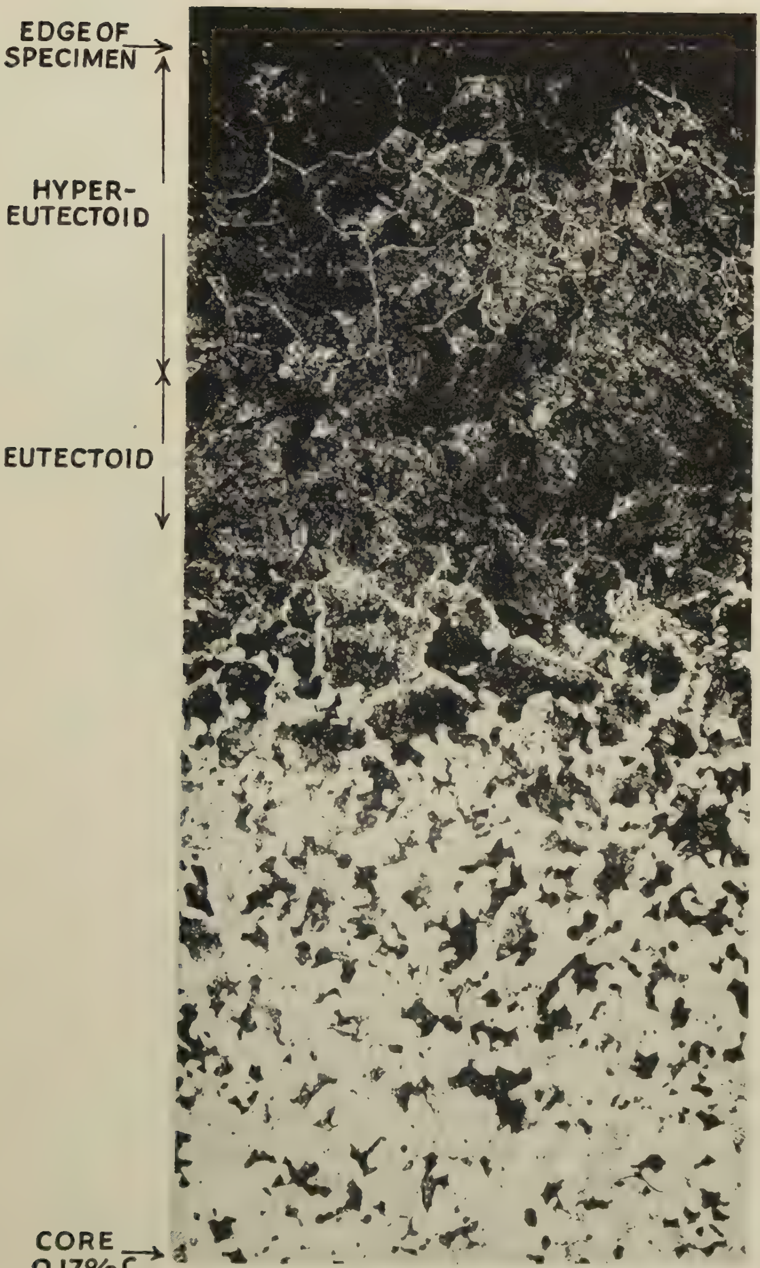

Fig. 4.—Gradation of structure in carburized zone of a 0.17% C steel. Box carburized 8 hr. at 1700°. Slow cooled. Structure normal, grain size 2 to 4 ×75 (Battelle.)

325,000, but it is so brittle that no one has been able to prepare pure cementite in test bar form. Were it so prepared it would actually show very low tensile strength, because so brittle a material cannot be uniformly loaded in tension, but it should have high compressive strength. Its ordinary occurrence in slowly cooled steels (with C greater than 0.85%) is either as a network, such as we have seen, or as spines and needles.

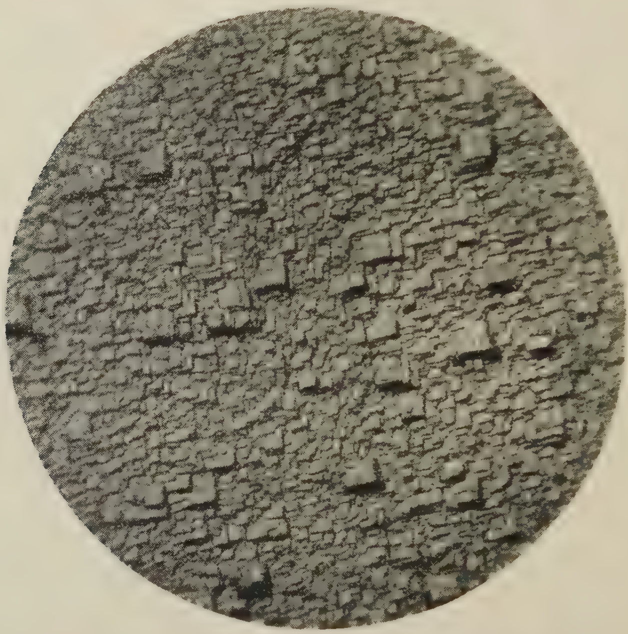

Fig. 5.—Slowly cooled iron, deeply etched to show crystalline structure. Etched in 9% HNO_3. × 900. (Dept. of Metallography, Univ. of Minn.)

Static Strength. We may now sum up these facts in their relation to the static * strength of slowly cooled (annealed) steel as follows: Free ferrite has a minimum tensile strength with high ductility; pearlite has a maximum tensile strength with low ductility; free cementite confers added hardness and a tendency toward brittleness, the steel tends to break apart along the network or the spines of cementite, hence in hypereutectoid steels with such structure there is a consequent lowering of the tensile strength. By increasing the amount of pearlite in the steel, the static strength is increased but there is a corresponding decrease in the ductility. As an increase in the amount of pearlite necessarily means an increase in the amount of C, the effect of increased C will be the same. This is shown schematically in Fig. 2, which shows that the strength of annealed steels reaches a practical maximum at a composition only slightly lower in C than that corresponding to the maximum amount of pearlite.

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* "Static" tests are tests made under a single, gradually applied load. Tests made under rapidly fluctuating loads or with a load rapidly applied are termed "dynamic."

Heat Treatment. Heat treatment in general consists in changing or regulating the structure of the steel by various methods of heating and cooling. By the term "structure" is meant (1) the nature and distribution of metallographic constituents, among which are those just described; (2) the size of the grain; (3) the network. In order to understand the nature of these changes and their application, it will be necessary to have a clear understanding of the mechanism by which they are brought about.

Critical Points. The structure of most steels may be modified or entirely changed by various heat treatments. These structural changes take place in the steel while it is in the solid condition, at temperatures known as the "critical points" or "critical temperatures" of the steel. For brevity, one often omits "points" or "temperature" and speaks of the "critical." These critical points are denoted by the letter "A," for "arrest," followed by the letter "c" (abbreviation for the French word "chauffage" (signifying "heating," the "c" can be recalled by thinking of "calorific"), or the letter "r" (French, refroidissement," signifying "cooling"; the letter "r" can be recalled by thinking of "refrigeration"). These signs, Ac or Ar, are further modified by the subscripts $_1$, $_2$, $_3$, indicating the particular point referred to. Thus Ac_1 means the first critical passed upon heating the steel beyond a certain temperature, and so forth. These critical points are indicated graphically in Fig. 6. This diagram refers to heating, not to cooling, as the critical points are somewhat depressed on cooling, the amount varying with the rate of cooling. The critical points on heating, for commercial C steels, differ slightly, and for alloy steels very considerably, from those shown for pure Fe-C alloys. When equilibrium conditions of extremely slow cooling as well as slow heating are being discussed, it is considered the Ac and Ar temperatures would be the same, in which case the notation A_1, A_2, A_3, without the letters c or r, may be used.

It should be here emphasized that Fig. 6 represents an "equilibrium" diagram, and that the lines on it represent extremely slow rates of heating, much slower than are encountered commercially.

First, let us consider the case of very slow heating of a 0.20% C steel, slow enough so that equilibrium conditions are nearly present at all times. Practically no change in the constituents occurs during heating until a temperature corresponding to the lower critical Ac_1 is reached. This begins about 1335° to 1340°. In passing through the lower critical there is a complete change in the nature and structure of the *pearlite,* it being converted into an entirely new constituent with new characteristics. This technically is known under the general term of a "solid solution," micrographically called "austenite.*" The *excess* ferrite remains unchanged.

Solid Solutions. To understand better the nature of this new component consider the interaction between salt and ice. When these two substances are placed in contact with each other under suitable conditions of temperature, the salt and ice merge into one another and pass from the state of two separate substances, a mechanical mixture, into that of a single substance, a brine solution. A similar process takes place in the case of the pearlite, except that the resultant solution is *solid* instead of liquid. The individual plates of ferrite and cementite which have characterized the pearlite grains now merge into one another, forming this new substance or constituent, a solid solution, which, save that it is a solid and not a liquid, has much the properties of a liquid solution. Its original components are merged into a single entity, giving a complete indefiniteness of composition, and with entirely new characteristics.

Absorptive Power of Austenite. Just as the brine solution can dissolve more salt or ice with increased temperature, so this solid solution of iron and iron carbide possesses the power of absorbing or dissolving more free ferrite or free cementite. Therefore, as the temperature is raised above that of the lower critical (Ac_1), there being an excess of ferrite in this particular steel (0.2% C), the solid solution or austenite begins to absorb the ferrite. This continues progressively with increased temperature until the upper critical, Ac_3, is passed. In passing through the range, from Ac_1 to Ac_3, which ends at about 1565° to 1570°, the last of the remaining excess of ferrite is absorbed by the austenite, so that above the upper critical, the steel is composed entirely of austenite—the solid solution.

The term austenite does not connote anything as to percentage composition any more than does the term brine. Austenite can hold in solution anything from 0 to 1.7% C, depending upon temperature. The limits of temperature and composition are shown by the area 4 of Fig. 6.

For a more rapid rate of heating than has just been considered, conditions will be somewhat altered. As the temperature rises above

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* In honor of an early metallographer named Austen.

Ac_1, the pearlite will rapidly change to austenite. This transformation will not be complete until the temperature is higher than Ac_1; how much higher depends on how fast the steel is heated. As the temperature rises towards Ac_3, the grains of austenite absorb those of ferrite by a process of reaction and diffusion, until all the ferrite has been changed to austenite. This process will not be completed until considerably above the temperature of the Ac_3 point, again depending on rate of heating. At the birth of the first austenite to form in our 0.20% C steel the newly born austenite has not 0.20% but 0.85% C.

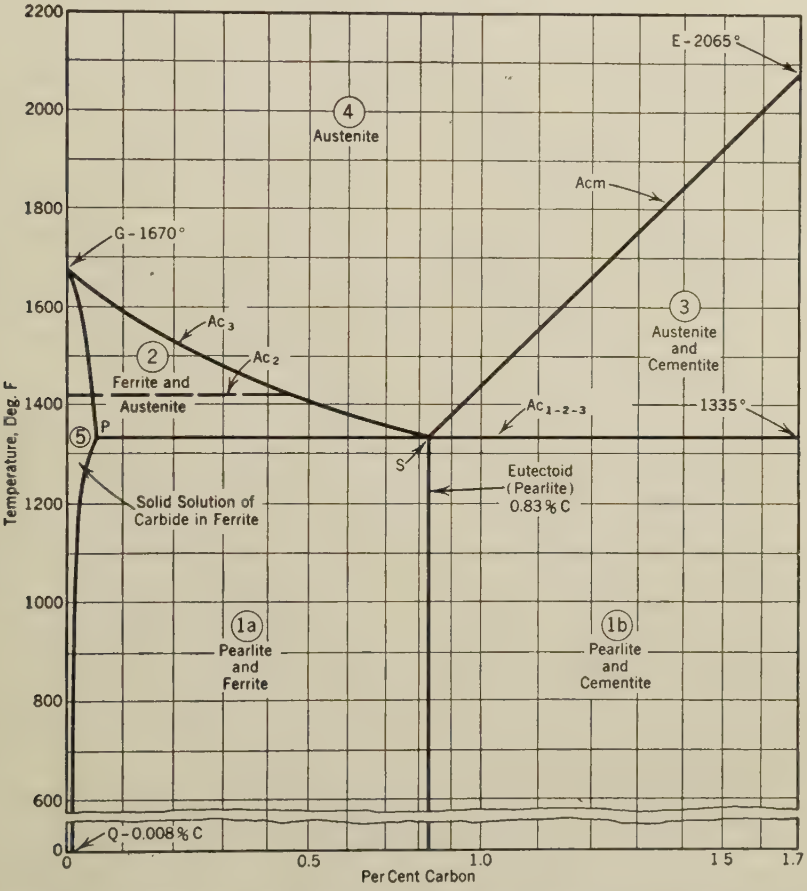


FIG. 6.—Iron-Carbon equilibrium diagram, showing changes on heating. Ac_1 occurs at the line PS. The lines GS and SE, bounding the austenite field are of primary importance in the heating phase of heat treatment, line PS in the cooling phase.

The grains are "cored," like many "cored" solid solutions familiar to the nonferrous metallurgist.

However, if the heating is stopped above the Ac_3 temperature but before all the ferrite is changed to austenite, and the specimen is held at that temperature, the austenite grains will continue to absorb the ferrite until all of it has been absorbed. At this moment the austenite does not have a uniform composition. Some parts of the austenite grains have a much higher C concentration than others. If sufficient time is given and the temperature held above the Ac_3, the C diffuses away from the points of high concentration to those of low concentration until finally the composition of the austenite everywhere is the same.

It is not known with certainty whether austenite holds C in solid solution as the element or whether it holds it as cementite. One may either think of the action as the straight solution of cementite, or may postulate that cementite, Fe_3C, decomposes into Fe and C before going into solution. However, since chemical analysis reports the content of C as C rather than as cementite, one generally speaks of the C content of austenite rather than of its cementite content, irrespective of one's views of the mechanism of solution. It makes no difference in our discussion of heat treatment whether one or the other mechanism is postulated. The changes on heating are illustrated graphically in Fig. 7. It will be seen that the initial pearlite, comprising about 25% of the normal steel, changes into austenite (the solid solution) at a temperature corresponding to that of the lower critical, Ac_1, and then progressively absorbs the free ferrite until, at a temperature corre-

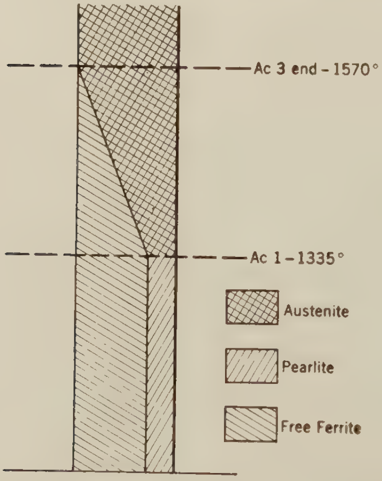
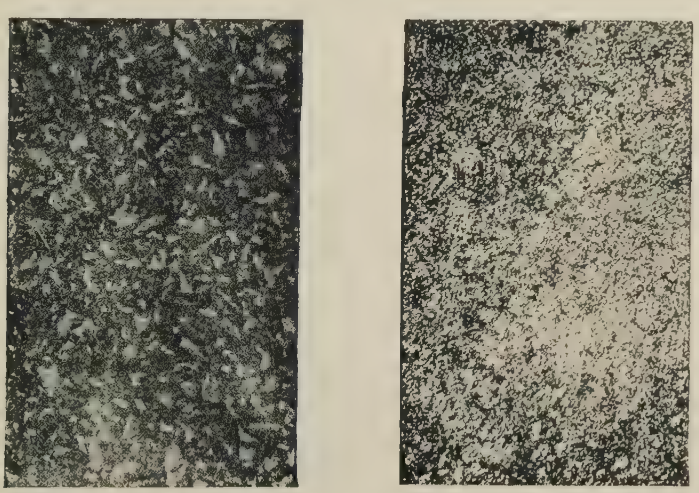


FIG. 7.—Change of pearlite and free ferrite into austenite during heating of 0.20% C steel.

sponding to the end of the critical range, Ac_3, the steel consists entirely of austenite.

The difference between heating past Ac_3 and heating only into the range between Ac_1 and Ac_3 is shown in Fig. 8. The micrograph on the right is of a low-C steel heated slightly above Ac_3 and quenched. That on the left is of the same steel heated only about halfway between Ac_1 and Ac_3 and then quenched. The light patches in the one on the left are ferrite that was not absorbed by austenite. (The ground mass is not austenite after it is quenched but its decomposition products, which will be considered hereafter. Quenching does not affect the ferrite, and so retains the unabsorbed ferrite in the location it had in the austenite.) Obviously, when we are heating to bring about the formation of austenite, we do not want to stop short of fully accomplishing our end. With very rare exceptions, in heat treatment depending on the formation of austenite, it is essential that the transformation be carried to completion, so *one of the most important precautions in heating for annealing, normalizing, or quenching, is to be sure that* Ac_3 *has been passed.* The important exception is in spheroidizing where annealing is done just below or started from just above Ac_1. However, such treatment is always preceded by one in which the steel has been taken above Ac_3.



(*a*) Quenched from between Ac_1 and Ac_3. Non-uniform structure.

(*b*) Quenched from above Ac_3. Uniform structure. × 100 (Bullens.)

FIG. 8.—Improper and proper quenching of low-carbon steel.

The So-called "Beta" Modification of Ferrite. Associated with these critical changes there is also a change in the ferrite (iron). Ferrite in its normal condition is called alpha ferrite, or alpha iron, and is characterized by ductility and by possessing magnetic properties. Upon heating this alpha ferrite to a little over 1400°, corresponding to the critical point Ac_2 in Fig. 6, the iron becomes practically nonmagnetic and then is called beta ferrite or beta iron.

Only in respect to magnetic properties is it necessary to consider any differentiation between alpha and beta. In heat treatment we can entirely disregard the controversial question of what differences may exist and can proceed just as though there were no magnetic change. The only reason for showing it at all in Fig. 6 is to indicate why Ac_3 is given the subscript 3 instead of 2.

Gamma Iron. Upon further heating to a temperature above the upper critical, Ac_3, there is a change in the crystal structure of the iron, it being known then as gamma iron. This is an allotropic change.* Gamma Fe has the property of being able to dissolve large amounts of C or iron carbide, while alpha Fe can hold only about 0.04% C as a maximum at Ac_1 and only about 0.008% at room temperature.

The differences between alpha and gamma Fe are due to the configuration of the atoms in the iron crystal. Alpha Fe has the atoms arranged as shown in Fig. 9-1. This structure is called "body centered cubic" from the position of the atom within the framework formed by the other atoms, whereas gamma Fe has the structure of Fig. 9-2, "face centered cubic." Both types of crystals are in the cubic system. This information is gained by X-ray crystal structure methods of study. The terms body centered and face centered are common in modern metallurgical literature. Since in ordinary steels one brings the metal into the austenitic condition by heating in order to cool it at a controlled rate, one can remember that austenite is *f*ace centered since it is made ready *f*or cooling. Another way is to recall that the face is above the body, and austenite is above ferrite in the Fe-C diagram.

Merging of the Critical Points. By reference to the diagram in Fig. 6 it will be noted that at the eutectoid ratio of C, that is, at about 0.85% C, the three critical temperatures, Ac_1, Ac_2, and Ac_3, merge into one. That is, steels consisting of pearlite alone, when heated to a temperature beyond this point, will change directly into the solid solution, i.e., the steel will consist of austenite (C or carbide) dissolved in gamma Fe. Similarly, as normal pearlite represents this eutectoid ratio, the same change of pearlite into a solid solution of carbide in gamma Fe will always occur at this temperature in ordinary C steels irrespective of the C content of the steel as a whole.

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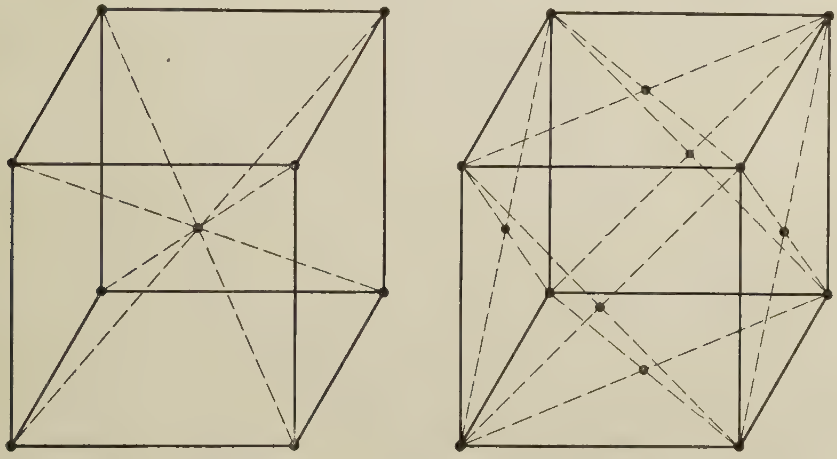
* Sauveur defines allotropy as "suggesting marked and sudden changes in some of the properties of a substance occurring at certain critical temperatures, without any change of state or of chemical composition."

Changes on Heating Different Steels. With this explanation in mind, we may now refer to the example of the 0.2% C steel and explain more fully the changes which take place in the constituents. Under normal conditions, this steel will consist of pearlite plus alpha ferrite. Upon heating through Ac_1, the pearlite will change into austenite, the Fe of which will be in the gamma modification; the free ferrite will still remain in the alpha condition. Upon further heating above Ac_1, the austenite will begin to absorb the free ferrite. (Upon passing through Ac_2, the steel will become non-magnetic.) Upon further heating within zone 2 the remnant of the ferrite will be gradually absorbed, so that on passing through Ac_3, the whole steel will be austenitic, a solution of Fe_3C (or C) in gamma Fe.



(1)
Alpha, body centered.

(2)
Gamma, face centered.

FIG. 9.—Arrangement of atoms in the allotropic forms of iron.

In a similar manner we might explain the changes in constituents which take place upon heating normal steel with any C up to that of the eutectoid ratio. With a C content approximately 0.5% the Ac_2 and Ac_3 points merge into one, known as $Ac_{2\text{-}3}$.

In a manner analogous to the absorption of free ferrite by the solid solution in the hypoeutectoid steels, the free cementite will be absorbed in the case of the hypereutectoid steels, the final solution taking place at a temperature range indicated by the line Acm. The only difference, and that a practical one, is that the solution of the free cementite takes place more sluggishly than the solution of the free ferrite of the lower-C steels.

Heat treatment of hypereutectoid steels does not always require that the last traces of cementite be taken into solution. Ac_{1-2-3} must always be exceeded, and a temperature either above or rather near Acm usually needs to be reached.

Determination of Critical Temperatures. "Thermal Analysis." A knowledge of the critical temperature on heating, especially of the end of Ac_3, is necessary in heat treatment, since this temperature must be exceeded in heating for full annealing, normalizing, or quenching. The temperature can be determined by quenching specimens of the steel from several temperatures around the expected Ac_3 temperature and determining which ones show the full hardening possible with a steel of the type being studied. The approximate temperatures for the S.A.E. series of steels are given in the "A.S.M. Handbook," * also in *Metal Progress.* † Variations from these temperatures due to variations in C and alloying elements within the permissible range of composition for a S.A.E. steel may be of importance when the heat treater desires to go above but keep very near Ac_3. Approximate corrections can be made if the chemical analysis is known.

The usual methods of determination require the plotting of heating curves of temperature versus time or of temperature versus temperature difference between the steel and a transformationless "neutral body," and noting the "arrests," i.e., the changes in the curves due to absorption or evolution of heat. These methods are briefly discussed in the "A.S.M. Handbook." ‡ Various ways of plotting data and for refining the method for very precise determinations are discussed in an A.S.T.M. publication. §

The Ar Changes. Corresponding critical changes take place upon cooling slowly from above the upper critical, except that they occur in the reverse order and with opposite effect. On account of the atomic inertia in solid metals and alloys, however, we find that in cooling these criticals Ar_3, Ar_2, Ar_1 are below the temperatures at which they appeared on heating. This difference is dependent upon length of exposure and the temperature to which the steel has been subjected, the rate of cooling, and more particularly, upon the influence of the alloying elements which may have been added to the steel. Some of the alloys, if present in sufficient amount, will cause the critical points to fall very low, even below normal temperatures, i.e., to produce air-hardening steels. The amounts by which Ar_3 and Ar_1 are lower than Ac_3 and Ac_1 in commercial steels depend upon the composition of the steel and the rate of cooling. For ordinary C steels in furnace cooling or the cooling of fairly heavy pieces in still air, the differences are of the order shown in Fig. 10.

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* 1937 Ed., p. 361.

† October, 1937, p. 462.

‡ 1937 Ed., p. 506.

§ Recommended Practice E14–33. See also articles in the bibliography appended thereto.

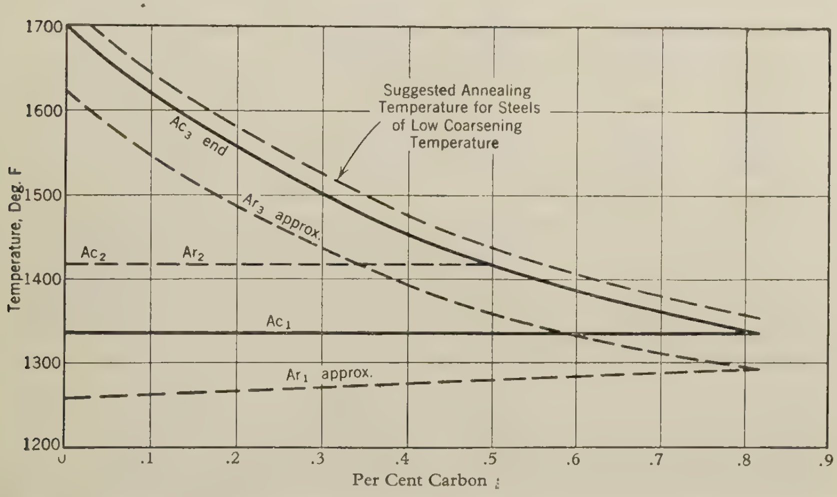

FIG. 10.—Relation of critical range on heating and cooling to temperature for full annealing of hypoeutectoid C steels.

During the cooling that takes place in forging or rolling, the steel remains austenitic, till Ar_3 is reached. That Ar_3 is at a lower temperature is fortunate since it increases the temperature range for hot working. As we shall see later, that the possibility exists of still further depressing Ar_3 and Ar_1 by more rapid cooling is another most fortunate circumstance.

Ar_2, the magnetic change, is not materially depressed on cooling in C steels. This is one of the ways in which this change differs from the true allotropic changes. By suitable alloying, Ar_2 may be depressed, but even in the alloy steels, whatever its position is, it is ordinarily not materially affected by changes in rate of cooling.

Changes on Slow Cooling. Upon slow cooling from above the upper critical, the solid solution will commence to reject the excess ferrite (or, of course, the excess cementite in the case of hypereutectoid steels) as the temperature decreases from Ar_3 or Acm to Ar_1. The reverse changes in the physical nature and properties of the Fe from those previously noted under heating occur at the criticals during cooling. When the lower critical is reached, the excess ferrite or cementite will have been entirely rejected, and as the steel passes downward through this point the solid solution—now containing 0.85% C—will change into pearlite. Under similar conditions of cooling, the original steel and the present heated and cooled steel will have the same structure.

Refinement. Before leaving the subject of the influence which heating through these various critical ranges has upon the structure of the steel, there are a few points concerning refinement which should be mentioned briefly. Again assuming that the steel is in the normal condition, no change will take place in the structure until the temperature has been raised at least to that of the lower critical.* At this temperature the original pearlite grains are completely changed and refined. If the steel has a C content other than that of the eutectoid ratio (i.e., contains free ferrite or free cementite), the *steel as a whole will not be refined;* the excess ferrite or cementite will remain unaltered and the steel will retain practically its original grain size. Complete refinement of the steel as a whole will not result until the steel has been heated to a temperature slightly above the upper critical. A clear understanding of these principles must be had, since they form the basis of many of the heat-treatment processes which will be developed later.

Grain Size beyond Ac_3. As the temperature is progressively raised above the upper critical, a gradual coarsening of the austenite grains occurs. This increased size is not only a function of the temperature, but also of the length of time at which the high temperature is maintained, and, is dependent as well upon certain characteristics within the steel itself, which will be discussed in Chap. 9.

Until we reach that chapter, our discussion refers to steels with so-called "inherent coarse grain," but this discussion will have to be modified later in order to apply to the controlled grain-size steels which are inherently resistant to coarsening.

Network. The third factor in the structural changes taking place upon heating is the effect of temperature upon the network. All hypoeutectoid steels in the normal condition are made up of pearlite with a varying amount of excess ferrite, the latter decreasing with the increase in C. From our study of the internal mechanism by which the constituents of the steel are formed by slow cooling, we know that the pearlite forms last, the ferrite being rejected by the solid solution ahead of the pearlite. Being thrown out to the boundaries of these austenitic grains, the excess ferrite forms a network around them. Upon reheating, this network is gradually absorbed, its final absorption taking place upon passing the upper critical. This change is similar to that explained previously under the description of the action of the excess ferrite.

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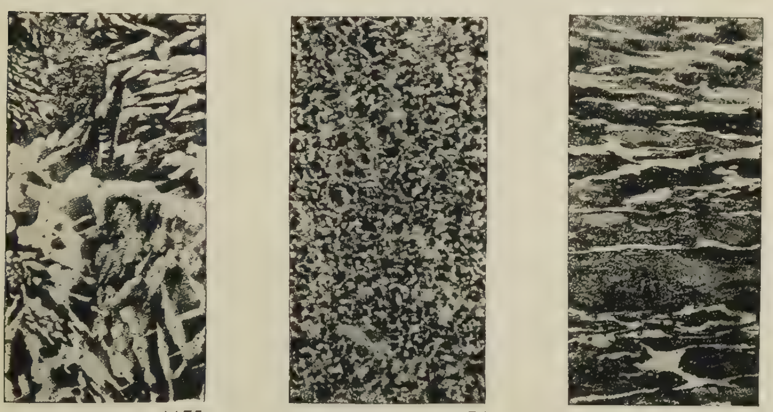
* Very long sojourn just below the lower critical does, however, result in spheroidization, as discussed in Chap. 4.



(Ordnance Dept.)

	(*a*)	(*b*)	(*c*)
Tensile	77,000	83,500	....
Yield	39,000	50,500	....
Elong. %	10.5	27.5	....
R.A. %	17	43	....

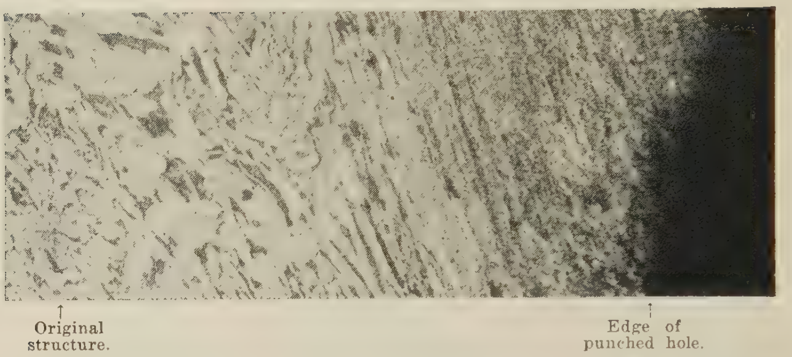
(*a*) As cast, coarse structure.

(*b*) Hot forged, finished at 1450°, fine structure.

(*c*) Cold worked, grains distorted.

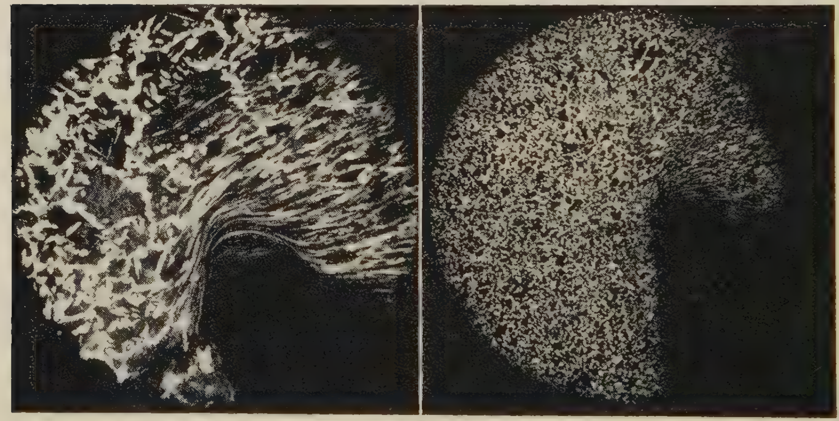
FIG. 11.—Structure of steel in different conditions of working.

The Effect of Work on Grain Size. Steel cooled slowly and undisturbed from a sufficiently high temperature will show a coarsely granular or crystalline structure, and the size of the grain is a function of the temperature and time during which the material is held at the maximum temperature, and the rate at which the material is cooled. In large masses the structure will be coarser in the center than at the



×100 (Epstein, Battelle.)

FIG. 12.—Distorted structure around cold-punched hole in structural steel angle ½ in. thick. This brittle, distorted, structure extends for about 0.02 in. back of the periphery of the hole.



(a) (b)

× 235 (Wysor, Bethlehem Steel Co.)

FIG. 13.—Cold-worked structure obtained in machining a 0.50% C axle. The cut was interrupted and the tool backed away from the un-completed cut. The surface of the completed cut and the part ahead of the nose of the tool which will form the chip, are badly distorted in the coarse grained specimen (*a*), which is in the as-forged condition, 149 Brinell. The surface and chip of (*b*), which is the same steel in the normalized condition, 163 Brinell, is not so badly distorted, owing to the finer grain and greater hardness.

surface, owing to the difference in rate of cooling. In order to overcome this difference and at the same time produce a homogeneous, uniform material, the steel is worked during the period at which grain growth ordinarily would take place. Steel that has been hot worked down to the Ar_1 point will show a finer grain, and will be stronger than the same steel slowly cooled without work, and at the same time will show high ductility.

The direction of the grain is of major importance in forgings and materially affects ductility in rolled material (compare paragraph on directional ductility, Chap. 1). Both the distribution of the work, controlling the grain direction, and the finishing temperature, controlling the grain size, affect the final properties. Examples of steel as cast and after forging are shown in the micrographs of Fig. 11, *a* and *b*.

Steel that has been worked below Ar_1—that is, cold worked—will show considerable distortion of grain, as is illustrated by Fig. 11*c*. Rolling too cold frequently develops a weak, laminated structure. Intentional cold work is usually performed at room temperature on steel previously cooled, rather than as the temperature falls below Ar_1 at the end of a hot working operation. Cold rolling of sheet and strip or cold drawing of rods and wire greatly strengthens and hardens the steel. Extreme cold deformation brings the steel to such a degree of brittleness that it must be softened by annealing before further deformation can be given without causing cracking. Partial softening by regulated annealing may be used to strike a happy mean between the conditions of extreme hardening by cold work and extreme softening by full annealing. When the properties aimed at are obtained by a regulated amount of cold work, a low temperature anneal is often required for release of internal stress, as discussed in the next chapter.

Even punching or machining operations may greatly affect the structure as shown in Figs. 12 and 13. The embrittled skin so produced may have to be removed, as by reaming a punched hole, a topic which has been discussed by Epstein,[1] or by following a rough machining operation that produces a worked skin, with a fine cut.

BIBLIOGRAPHY

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CHAPTER 3

GENERAL ANNEALING PRACTICE

The properties of steel are primarily controlled by its structure, which may be mechanically altered by cold work, or thermally altered by annealing, normalizing, or quenching and tempering. Annealing usually aims chiefly to improve ductility with a decrease in strength, normalizing to produce reasonable ductility without great sacrifice in strength, and quenching and tempering to produce some combination of both strength and ductility aimed to meet definite requirements. When these processes have reached their limit as applied to C steels, alloying the steel opens up a still wider range of properties produceable by these processes. The processes themselves are best discussed in relation to C steels, and in the order mentioned above.

Annealing. Annealing, in its commercial application, may have for its purpose any or all of the following aims: (1) to "soften" the steel and thus put it in condition for machining or to meet certain physical specifications; (2) to relieve internal stresses caused by previous hardening or elaborating operations; (3) to obtain refinement of the grain in combination with high ductility.

Thus, depending upon the results desired, *commercial or process annealing* will consist of a heating operation carried to some predetermined temperature—which may be below the critical—to produce the results desired in items 1 and 2, and followed by a moderately slow cooling of the metal from that temperature. *True or full annealing* for item 3 requires heating to above the upper critical, followed by slow cooling, usually in the furnace.

Heat Application. In the abstract, annealing would appear to be but a suitable correlation of the elements of temperature, time, mass, and surface. But in actual practice the success to be obtained in annealing (or in any heat-treatment process, for that matter) depends not only upon an understanding of the technique of the process, but also upon the judgment and skill of the furnace operator in applying the basic principles which may be derived from a consideration of the above factors. The manner of placing the charge in the furnace, of regulating the flow and composition of the hot gases, of determining when the steel has been heated uniformly and thoroughly, all matters of personal judgment, vitally affect the results. Particular stress should be laid upon the necessity for getting heat to the center and bottom of the charge, not only for the sake of uniform annealing, but also to shorten the time of saturation and lessen the time of exposure of the top and outside edges of the charge to the heat and influences of the chamber atmosphere.

A uniform chamber temperature does not mean necessarily a uniformly heated product; a circulation of heat through the mass is more desirable than the mere application from the outside; with the same chamber uniformity, it is possible to vary the quality of the anneal by the manner in which the steel is placed in the heating zone. It is advisable to raise the charge above the furnace floor or hearth upon suitable blocks or supports, to separate each piece from the other, and to avoid localized heating due to overloading. It is only by such means that there will be provided an opportunity for the circulation of the hot gases through the charge. In furnaces heated by electric resistor elements or by gas-fired radiant tubes, where there is no circulation of gases, the arrangement of the work so that it may uniformly receive radiant heat is of analogous importance. (See Chap. 16.)

Elementary Considerations. Fundamentally, an annealing operation consists of two well-defined and coordinated phases, the heating phase and the cooling phase. Just as a photographer must first expose his plate to the chemical action of light, properly time his exposure, and then by other chemical means develop and fix the image on the plate, so must the annealer first expose his steel to the action of heat and for a reasonable time, in order definitely to impress upon it characteristics capable of being developed and brought to completion in the second or cooling phase.

Whether the phase being considered is one of heating or of cooling, of heat absorption or of heat emission, it is concerned with a suitable correlation of the basic factors of temperature and time.

The temperature to which the steel is to be heated is that at which the desired change in grain size, microconstituents, or both will take place. The necessary rate of cooling is controlled by the characteristics of the steel.

Aside from the general question of initial heating, the time element in the heating phase is one of duration at the predetermined temperature, i.e., of the *saturation* necessary, and is dependent metallurgically upon the microstructure of the steel as received for annealing, the mass, and the surface.

"Saturation" requires the homogenization of the austenite in respect to C and other elements dissolved in it, and the attainment of temperature uniformity throughout the mass.

The time element in the cooling phase is an expression of the *rate* at which the steel shall be cooled through and below the critical and is determined by the microstructure desired in the steel, the mass, and the surface.

Initial Heating. Slow, careful, and uniform heating always is advisable, regardless of the chemical composition or physical condition of the steel. Heating to such temperatures as are common in general annealing practice necessarily results in more or less change of physical condition or molecular readjustment; and the greater the hardness, brittleness, and amount of internal stress in the metal, the greater will be the deleterious effect of too rapid heating. Thus, objects of intricate design, or with varying cross sections, or steel in a hard, brittle condition, should be given the greatest care in heating in order that the release of stress may not cause warpage or otherwise injure the metal. Such pieces should never be placed directly in a hot furnace, but should be given a careful preheating. Steels with low thermal conductivity require especially slow and careful preheating.

Microscopic Changes. In the previous chapter it was explained that, in the ordinary cast, hot-rolled or forged sections (pearlitic in character), there is virtually no change in grain size or in constituents during the heating to a temperature below the lower critical, Ac_1. That is, there is no refinement of the steel.

As the temperature passes Ac_1 there occurs the complete change of the pearlite to the solid solution, giving the maximum refinement to the austenite.

With increasing temperature the excess ferrite is absorbed progressively by the solid solution and is completed at or above Ac_3. If the rate of heating through this range of temperatures is comparatively slow, there will be a complete absorption of the remaining ferrite at Ac_3.

The Upper Critical. As the steel passes through Ac_1 multitudes of tiny austenite nuclei are formed. As it passes the upper end of the critical and the structure has become wholly austenitic, there is complete refining of the grain. As the temperature increases beyond this range the grain size tends to coarsen, and at a sufficiently high temperature, as will be discussed in Chap. 9, the coarsening may be extreme. When this occurs the effect upon the physical properties of the steel is great.

The behavior of 0.40% C steel (initially in a slightly cold-worked condition) upon annealing at different temperatures and slow cooling in the furnace is shown in Fig. 1. As the temperature rises, process annealing occurs, the steel softening gradually below Ac_1, and more rapidly thereafter, but not reaching maximum softness till Ac_3 has been passed.

At temperatures in excess of Ac_3 the yield strength of this particular steel rises for the first 200° and tensile strength is not materially affected. Ductility, however, drops sharply and falls off steadily with temperature above Ac_3. This is a result of the coarsening of the structure. The best combination of strength and toughness is obtained by avoiding coarsening, i.e., by not heating too far past Ac_3.

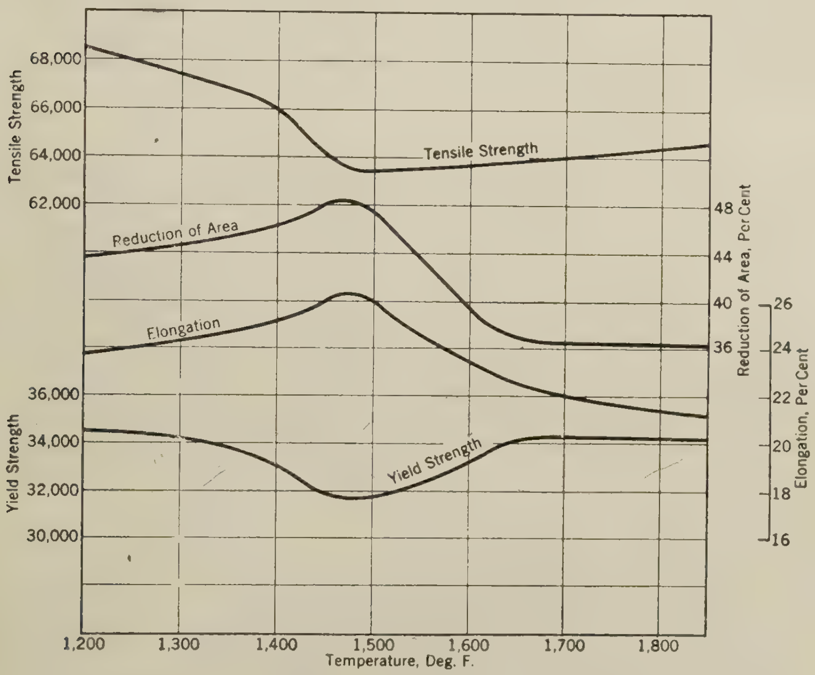
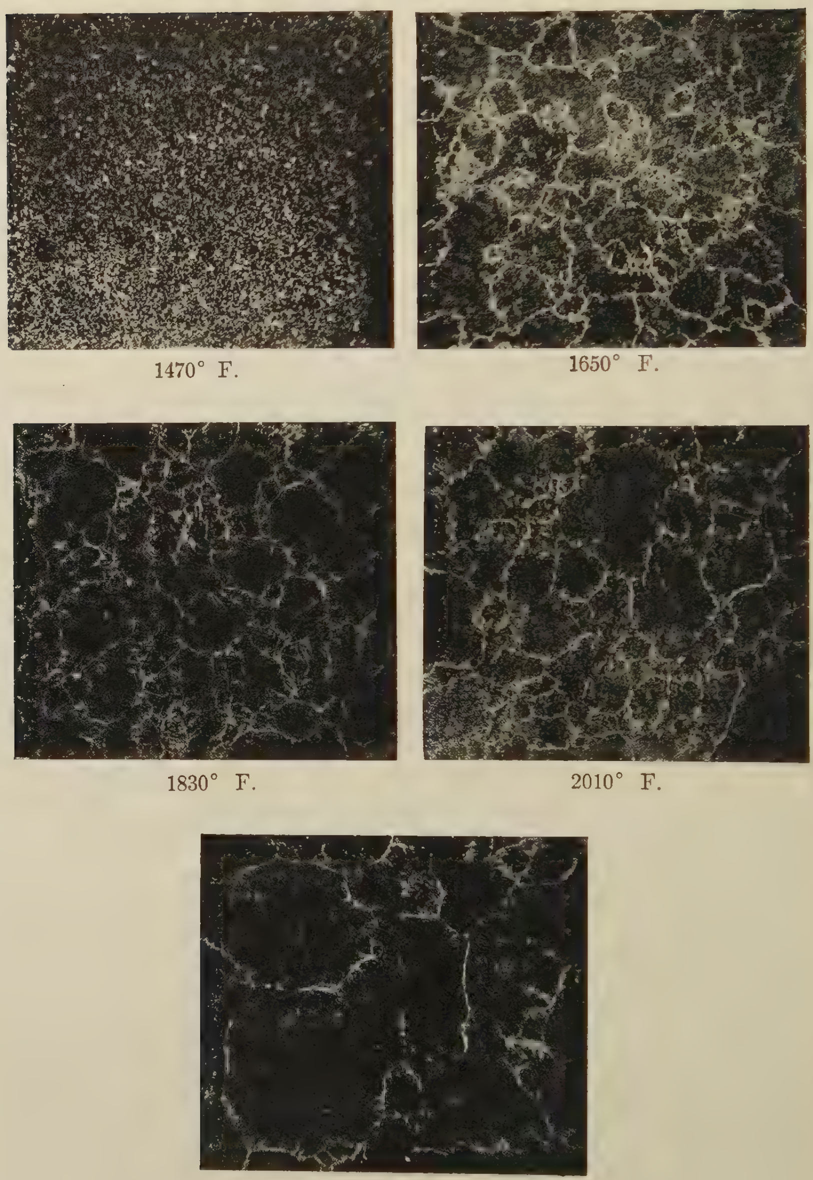


FIG. 1.—Effect of annealing temperature on mechanical properties.

A coarsened structure may be desired for other reasons, e.g., it machines better in roughing cuts, but such a structure has poor impact resistance. The loss of impact resistance goes on at a faster pace as the grain coarsens than does the loss of static ductility.

Figure 1 represents one particular steel, and the results with steels of other compositions and grain-coarsening characteristics will deviate



1470° F.

1650° F.

1830° F.

2010° F.

2190° F.

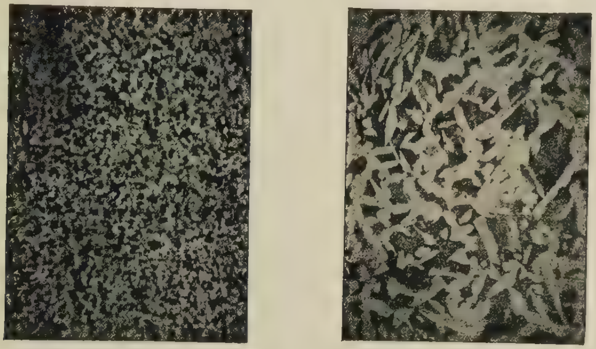
× 40. (Howe.)

Fig. 2.—Effect of heating beyond Ac_3. 0.40% C steel heated at temperatures indicated for ten minutes and air-cooled.

somewhat from the trends indicated. However, the trends shown are quite consistent with those given by Meyer and Wesseling[1] and reproduced by Sisco[2] for steels of 0.75–1.10% C. The Meyer and Wesseling curves also show the effect of long versus short holding time, which modifies the results.

The general conclusion is that, especially for long holding times and on steels readily subject to grain coarsening, it is not advisable greatly to exceed the upper critical temperature in annealing.

The grain size differences in the original austenite are shown in micrographs of normalized specimens in Fig. 2 from Howe for a 0.40% C steel that coarsens readily. Figure 3 shows the grain size of the same steel in the annealed condition at two temperatures.



(*a*) 1470°

(*b*) 2010°

× 40 (Howe.)

FIG. 3.—Effect of heating a 0.40% C steel for 10 min. at temperatures shown, followed by furnace cooling.

Temperature of Heating. Reduced to lowest terms, the true or full annealing operation requires the production of an entirely new crystalline structure, the constituents of which may be required to be of the smallest grain size attainable or of some required larger grain size; this operation also should eliminate all internal stresses. As previously described, this new structure is given birth at the Ac_3 temperature since only at or above this temperature can there be obtained a homogeneous austenite solution. The exact temperature will depend upon the chemical composition of the steel, chiefly upon the C. Practice differs in reporting the Ac_3 temperature, but it is often the custom to give the temperature corresponding to the maximum thermal change. It therefore is necessary to add some 35°—which has been shown by critical point determinations and quenching experiments to be the approximate difference—to the reported Ac_3 maximum temperature, in order to locate the end of the upper critical. A curve showing the Ac_3 end temperatures—and therefore the minimum full annealing temperatures—of hypoeutectoid steels has been given in Fig. 10, Chap. 2. The upper line in the same chart gives the temperatures recommended for steels of low coarsening temperatures. This allows a leeway of but 25°, which is sufficient, with the most careful methods of heat application and control.

With "controlled grain-size" steels (see Chap. 9), greater leeway is allowable and with many alloy steels (in which, of course, Ac_3 varies with composition) Ac_3 must be considerably exceeded.

In the "Standard Definitions" relating to heat treatment, phrased by a committee representing the American Society for Testing Materials, the Society of Automotive Engineers, and the American Society for Metals, it is stated that "the annealing temperature is generally about 100° above the upper limit of the critical range." This results in the following:

C Content, %	Range of Annealing Temperature
Less than 0.12	1600° to 1700°
0.12 to 0.29	1550° to 1600°
0.30 to 0.49	1500° to 1550°
0.50 to 1.00	1450° to 1500°

It will be noticed that the temperatures recommended by the Committee are distinctly higher—especially for the tool-steel grades—than the minimum values of the curve. The lower the temperature which can be used to give the desired results, the better, when dealing with steels of low coarsening temperatures, but this should not be interpreted as necessarily applying to other types of steel. It does not apply to heating for normalizing and may not apply to quenching.

Annealing Steel Castings. Notwithstanding what has been said above in relation to holding the temperature not far above Ac_3 in wrought steels, a much higher temperature and longer time at temperature are required when annealing steel *castings*. The original dendritic structure of the ingot has been mechanically broken up in rolling or forging so that in the heating of wrought steel the ferrite patches and the pearlite grains are intermingled and generally made much finer. The austenite grains with 0.85% C formed from the pearlite patches at Ac_1 do not have large ferrite boundaries and when Ac_3 is exceeded and the surrounding ferrite has also become austenite, but practically C free, the high-C centers and low-C borders of the austenite grains can equalize their C content fairly readily by diffusion. Moreover, the non-metallic materials rejected to the grain boundaries of the original ingot have also been distributed throughout the metal by the working.

In cast steels of the ordinary compositions, which have not had the benefit of working, the original coarse pearlite grains are surrounded by coarse ferrite envelopes, formed in cooling down after freezing, and the non-metallic boundary material has not been distributed. To absorb the coarse ferrite, break through the non-metallic films, and bring about a homogeneous C content in the austenite is a task requiring much higher temperatures and longer times.

Hence, whereas a *wrought* 0.30% C steel would preferably be heated only to around 1550° for annealing, a 0.30% C steel *casting* will require around 1650° and should be kept at temperature an hour for every inch of diameter or thickness in the largest section before starting the slow cooling.

Rate of Heating. Studying the heating phase from the practical aspect, there is another factor to be considered—that of bringing the whole mass of the steel to the proper temperature uniformly. It is self-evident that the center of a large mass of steel, such as a locomotive axle, shaft, or steel bloom, will lag in temperature behind the exterior. In other words, it is the tendency of the core to be considerably lower in temperature than the shell or outside of the steel. It is then a common procedure to raise the temperature of the furnace beyond the proper annealing heat in order to drive the heat to the center of the steel to be annealed. This is a great mistake. It is far better to take the extra time required to heat more slowly as the proper temperature is neared, thus bringing the steel to a uniform temperature throughout. If this is not done, the exterior of the piece will be carried beyond the proper temperature; and, in general, a needlessly high temperature is injurious and tends to coarsen the grain.

To express this matter in a different way, the furnace in which the metal is being heated for annealing should not be run at an indicated temperature higher than the maximum temperature to which the metal itself is to be heated. To illustrate: A piece of steel, such as a die block, heats, cools, and decarburizes on the corners first. The life of the entire piece of steel is no greater than the life of the corners. If the die block is placed in a furnace which is at the final working temperature, the corners are likely to reach the final temperature long before the major part of the mass. If the temperature is higher, as it frequently is as the result of a futile attempt to gain time, the corners are overheated before the center of the mass is heated and soaked to the proper degree. This example should indicate the necessity for slow, soaking heats in order to prevent overheating the corners of the metal. The necessity for prevention of oxidation or decarburization of the exposed edges is also obvious.

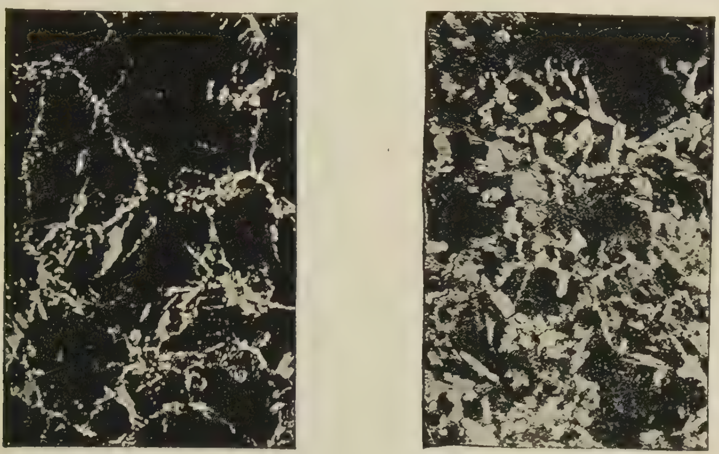
Time of Heating. The time element in annealing requires that the whole mass of the steel (1) shall be heated uniformly throughout at the proper temperature, and (2) shall be maintained at that temperature for a period sufficient to give birth to the new grain structure to relieve internal stresses, and give sufficient homogenization. Assuming a proper method of heat application, the duration of heating ordinarily is determined by the mass—the maximum size of section—and by the surface exposed to the heat. It will be shown later, however, that this time factor is also qualified most strongly by the initial structure of the steel.

For general purposes it may be assumed that, after the furnace has attained the indicated temperature desired, about one-half hour to one hour should be allowed for heating and saturation for every inch of section in wrought steel. This a very general statement and must be modified according to the experience of the individual operator and the methods of heat application.

In practice the heating rate is generally made much faster than the ideal rate, in order to secure greater production from a given furnace, and sometimes to reduce the amount of scaling. If the furnace atmosphere is controlled, scaling can be minimized or eliminated. A heating method which gives great speed plus freedom from irregular heating, but one applicable only to pieces of uniform cross section, is to pass an electric current through the piece and develop heat internally in the steel itself instead of letting it soak in from the outside at a rate limited by the ability of the steel to transmit heat. With such electric heating very rapid and very uniform heating can be accomplished and a higher temperature, leading to better homogeneity of austenite, can be utilized without allowing time for grain growth.

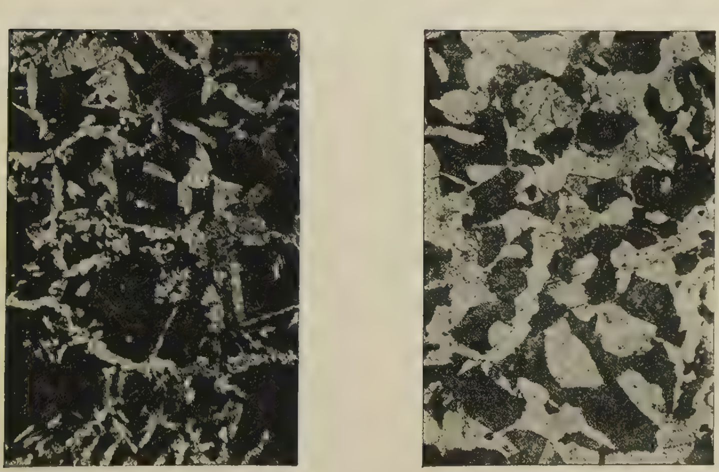
Behavior of the Excess Ferrite on Cooling. As previously explained, the cooling of a hypoeutectoid steel from a temperature above that of the upper critical to one below that of the lower critical will cause the austenite to reject the excess ferrite to the grain boundaries. The ordinary full-annealing operation generally will accomplish this separation of the excess ferrite, and if the cooling through this temperature range has been retarded sufficiently the excess ferrite will separate fully from the pearlite.

Thus, the cooling phase, depending upon the correlation of the elements of time, mass, and surface, will produce a structure varying from that of a fine network (*a* in Fig. 4)—indicating that very little of the excess ferrite has been separated because the rate of cooling was comparatively rapid; through the intermediary stages of coarse network (*b*) and partial separation (*c*); to that showing complete separation of the excess ferrite (*d*)—indicative of extremely slow cooling. The pieces of steel from which the micrographs were taken were of different sizes, from the same ingot, and annealed at a temperature



(*a*) Rapid cooling.
(Normalized).

(*b*) Slower cooling.



(*c*) Still slower cooling.

(*d*) Slowest cooling.
(Annealed)

×100 (Bullens).

Fig. 4.—Structure of a 0.40% C steel at different rates of cooling.

high enough to bring out more forcibly, by means of a comparatively large grain size, the definite structures desired; a finer grain could have been obtained by the use of a lower temperature.

With the same temperature of heating, if the elements of time-mass-surface in that heating phase have been adjusted correctly, it is evident that the time-mass-surface elements of the cooling phase, i.e., the cooling rate, will control the final structure of the annealed steel.

Cooling in the Furnace. Regardless of the size (mass-surface relations) of that piece, if the time element of cooling can be prolonged sufficiently, complete separation of the excess ferrite may always be obtained. From the practical standpoint the production of a structure showing complete separation of the excess ferrite always requires a cooling of the steel in the furnace—or its equivalent; and, further, that the temperature drop of that furnace (i.e., the temperature-time elements) during cooling must be in inverse proportion to the size of the piece (i.e., the mass-surface elements). This may be illustrated by the three following, widely divergent examples.

1. It would be good practice to assume that the commercial annealing of shafts 10 in. in diameter would require a large furnace of heavy construction, having a small radiation loss, and, therefore, a very slow temperature drop after the fires had been shut down and all doors, vents, and dampers closed. With such a method of cooling it would be immaterial whether the pieces to be annealed were 10 in. or 1 in. in diameter; in either case the steel would be cooled very slowly and complete separation of the excess ferrite would result. In this illustration the method of cooling controls the results, regardless of the size of the piece.

2. For the heating of a few pieces of 1-in. bars it might be considered good practice, under certain plant conditions, to use a small furnace of light construction, which would have the tendency to cool down rapidly when the fires were shut down. The cooling of small pieces in this furnace (which is assumed to have a radiation loss not in inverse proportion to the size of the steel) would not be retarded sufficiently to produce a state of complete separation of the excess ferrite, but only one of partial separation. In this case it is the mass-surface elements of the steel, as well as the methods of cooling, which indicate the final structure to be expected.

3. When the size, design, and construction of the annealing furnace are in proper adjustment with the mass and surface of the steel exposed to the heat, complete separation will result from slow-cooling the steel in and with the furnace, as is indicated by the structure resulting from the furnace cooling of a 4-in. bar, shown in Fig. 5*b*.

On account of the great variation in heat radiation during the cooling of steel with different mass-surface factors in furnaces of different size, design, and construction, the exact rate of cooling required to produce complete separation is one to be determined largely by experiment under the distinctive working conditions. The slower the cooling of the steel from the annealing temperature through the transformation range, the greater will be the degree of separation of the excess ferrite.

Other Slow-cooling Methods. If the objects are of large size, an approximation of furnace cooling may be obtained by removing the hot steel from the furnace and covering it with some blanketing substance of low heat conductivity, such as lime, ashes, or sand. In some instances, cooling small pieces in lime may result in a slower cooling than may be obtained in the furnace. When a large tonnage of steel must be annealed and there is not sufficient furnace capacity to permit of cooling in the furnace, a pit lined with brick or metal plates, and fitted with coverplates, may be used. By this method the hot steel is delivered immediately from the furnace to the pit and covered with ashes; cooling by this method of pit annealing often is slower than cooling in the furnace itself if the latter is not properly designed and constructed.

Structures Produced by Air Cooling (Normalizing). The structure that will be produced in steel with definite mass-surface factors, air cooled after proper soaking at a given temperature, may be determined with much exactness, because the relation between the heat radiation of the steel and the temperature of the air is definitely fixed and well known. Thus, the structure of Fig. 4*a* (a 1-in. bar) is characteristic of the air cooling of small pieces; Fig. 4*d* that of a large section. The fine network structure can be produced in large sections only with cooling by artificial means, such as quenching. The furnace cooling of small pieces, when not sufficiently retarded, often will produce results equivalent to the air cooling of large pieces. Normalizing will be more specifically dealt with in the next chapter.

Effect of Cooling Rate on Physical Properties. Ordinarily, a full annealing should produce the maximum refinement possible, such as is illustrated by Fig. 9*b*.

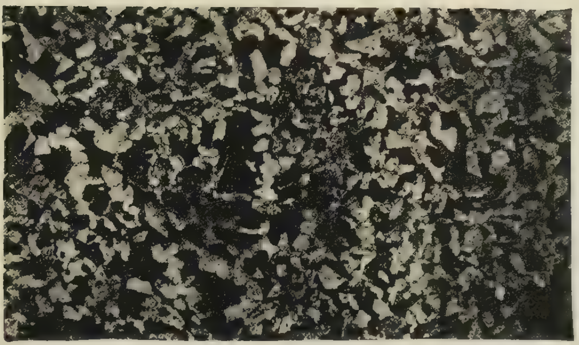
In order to illustrate the effect upon the physical results of varying the rate of cooling, a 12-in. square billet of 0.40% steel was forged into a 4-in. square bar. One piece of the bar was cooled slowly in the furnace from the annealing temperature, while another piece was cooled in the air. The annealing temperature selected was such as would produce a small grain size, and yet not so fine that the difference between the structures could not be recognized under the microscope at 100. The tensile test results are shown under Fig. 5, and indicate that the slower the cooling the lower will be the tensile strength.

Effect of Initial Structure. The previous discussion has been concerned mainly with the annealing and production of a definite structure in steel that has been subjected to proper heating and working in previous elaborating processes, such as rolling, forging, etc. Regardless of mass, if the steel to be full-annealed has a "reasonable" grain size before annealing, such as is shown in Fig. 5*a*, the time element will be normal, and an ordinary heating at the proper temperature will produce a small grain size. In other words, there should be no great difficulty in producing the desired results by annealing with an ordinary soaking at the proper temperature—a simple anneal.

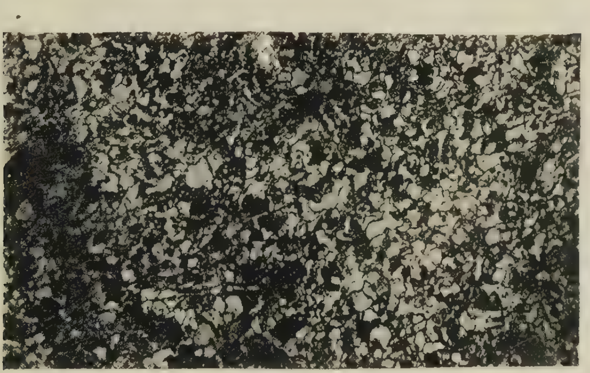
If such were the extent of the demands made upon the knowledge of the operator, the work would, with some experience, be comparatively easy. Unfortunately, however, existing plant conditions often make heat treatment a more complex problem, for such reasons as are given in Chap. 14 or as a result of some other elaborating operation, even though this has been properly performed. In cases in which the steel has been subjected to severe mechanical working, in large castings and forgings showing ingotism, or in large sections which previously have been overheated severely, the time of soaking may pass beyond normal limits and become a factor of magnified importance. Special annealing methods, or annealing involving several days of soaking, may even be required to bring about the necessary structural changes in the steel in the worst cases. The reasons for this may be explained by the slowness with which homogenization proceeds by the phenomena of diffusion and equalization.

Diffusion. Even though a totally new structure may be set up by the annealing heat, there remains for a considerable length of time a tendency of the new structure to return, upon slow cooling, to the arrangement in the original, even though the constituents themselves may be those born at the new temperature.

For an explanation of this, let us go back to our former simile of the salt and brine solution. When a grain of salt is dissolved by the brine, it is the solution in the immediate neighborhood of the salt crystal which acts as the solvent, and not the entire volume of the brine solution. In time, however, the dissolved salt eventually will diffuse through the whole body of brine and the brine then will be of uniform composition throughout. Now, a similar process is going on in the steel when the solid solution (austenite) is absorbing the excess



(*a*)—Forged from 12 in. to 4 in. square. × 100. (Bullens) Tensile, 78,600; yield, 32,500; Elong., 28%; R.A., 45%.



(*b*)—Annealed, Furnace Cooled. × 100. (Bullens) Tensile, 67,700; yield, 35,400; Elong., 35%; R.A. 55.5%.

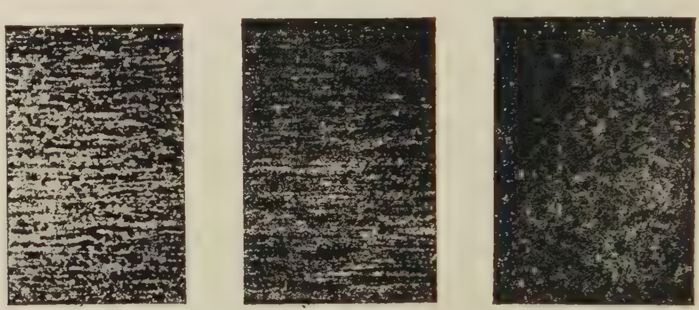
(*c*)—Normalized. × 100. (Bullens.) Tensile, 76,250; yield, 38,150; Elong., 34%; R.A. 56%.

When drawn at 1250° after normalizing the structure was not visibly altered. The properties were: Tensile 74,400; yield 33,000; Elong., 37.5%; R.A., 67.5%.

FIG. 5.—Effect of annealing and normalizing on physical properties. 0.40% C steel.

ferrite, and it will be found that *complete absorption may not mean complete diffusion or equalization.* The process of equalization takes time. The time is shortened with rise in temperature. If the passage through temperatures under that of the upper critical is slow enough, a large part of the diffusion will have occurred by the time Ac_3 is reached. In order that there may be complete homogenizing, and therefore, complete grain refining, the sojourn at a temperature above Ac_3 must be long enough for this complete diffusion of the absorbed excess ferrite.

This point is illustrated by Fig. 6, 5⁄32-in. thick NiCr steel plates: (*a*) shows the structure of the steel as rolled; (*b*) shows the steel after an ordinary annealing at a temperature somewhat above the upper critical; and (*c*) shows the same steel after a long anneal at the same temperature. It will be noticed that the steel in (*b*) has taken on approximately the same structural constituents as in the fully annealed piece, as shown in (*c*), but that it still remains in the arrangement of (*a*), even though the annealing temperatures were the same in both cases. The recurrence of reformation of these banded structures is due to the fact that complete effacement by equalization had not taken place. In other words, where these light-colored bands occur the C content, as a whole, is less than in the rest of the mass.



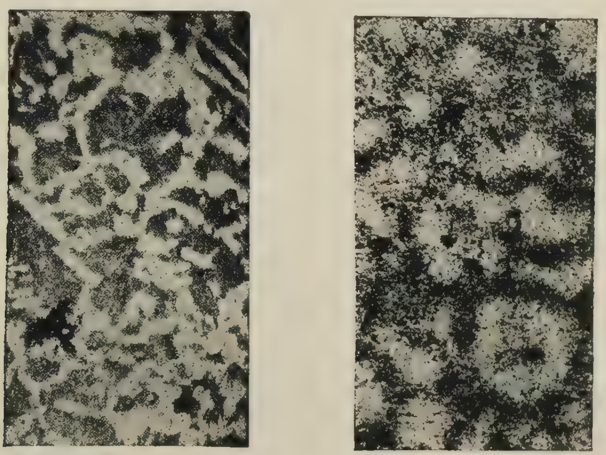
(*a*) As rolled. (*b*) Partly annealed. (*c*) Fully annealed.

× 60 (Bullens).

FIG. 6.—Progress of the annealing of a NiCr steel.

Special heating methods for securing diffusion and equalization are required, not only for the removal of the structures caused by severe mechanical working, but also for the elimination of the effect of severe overheating. This condition is illustrated by the example of a large forging which had been made under conditions of improper heating, and of insufficient working resulting from lack of proper forging equipment. After the forging had been put through a full annealing process at ordinary temperature and with normal opportunity for saturation, it was found by microscopic examination that little or no refinement had taken place. The annealing was repeated, but with similar results; the steel still showed the coarse grain of Fig. 7*a*.

A specimen was quenched into water from the annealing temperature, a little above Ac_3. The quenched structure of Fig. 7*b* reveals that the austenite had not been made homogeneous. It required a markedly higher temperature to produce satisfactory results. Thus the generally sound procedure of not materially overstepping Ac_3 was not applicable in this case.


(*a*) Slow cooled from annealing temperature.

(*b*) Quenched from annealing temperature.

×100 (Bullens).

FIG. 7.—Failure to refine on annealing, due to lack of homogenization of austenite.

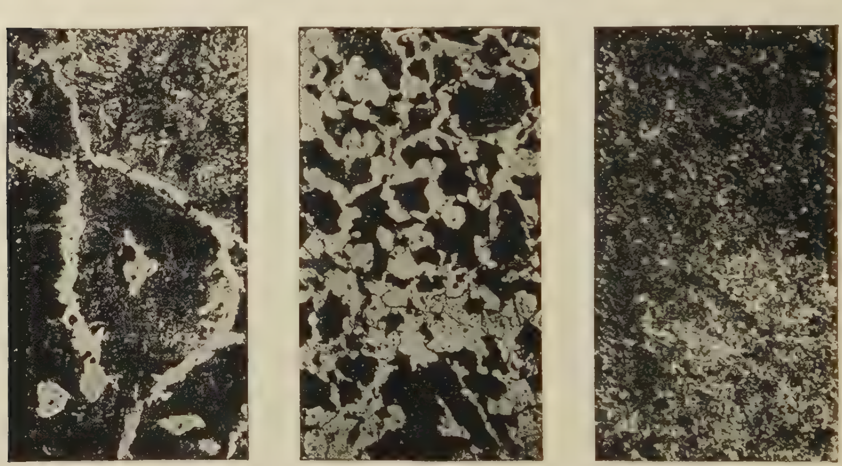
Refining Overheated Steel. Overheating in previous operations usually results in a coarse and irregular structure hard to break up.

The structure of an 8-in diameter 0.40% C steel shaft, which had been overheated, is shown in Fig. 8*a*. The tensile test results showed low ductility due to coarse grain. The tensile bar had a flat fracture.

Small pieces having an initial structure of this nature usually can be refined by a simple anneal, but with large pieces a double treatment may be required.

The micrographs, Figs. 8*b* and 8*c*, represent stages in the double treatment. The first heating at 1450°, followed by cooling in the furnace, served to break up the coarse grain (but without full refinement) and to redistribute the ferrite and pearlite; it also raised the ductility. A spheroidizing anneal, at 1400°, i.e., within the Ac_3-Ac_1 range, followed by cooling in the furnace, produced the structure shown in Fig. 8*c;* the tensile test gave a full cup fracture, and a very high ductility. The tensile properties in the three tests are shown in Fig. 8.

It is probable that a single anneal from a higher temperature with a very long soaking would have refined this steel without the necessity of the spheroidizing treatment, but a double heating and cooling are undoubtedly commercially quicker and produce greater refinement and higher ductility.



	(*a*) Initial, coarse, overheated structure, as forged.	(*b*) Annealed at 1450°.	(*c*) Annealed at 1450° then spheroidized at 1400°
Tensile	104,000	90,500	60,000
Yield	58,250	50,250	29,000
Elong. %	7	22.5	43
R.A. %	3	37	69

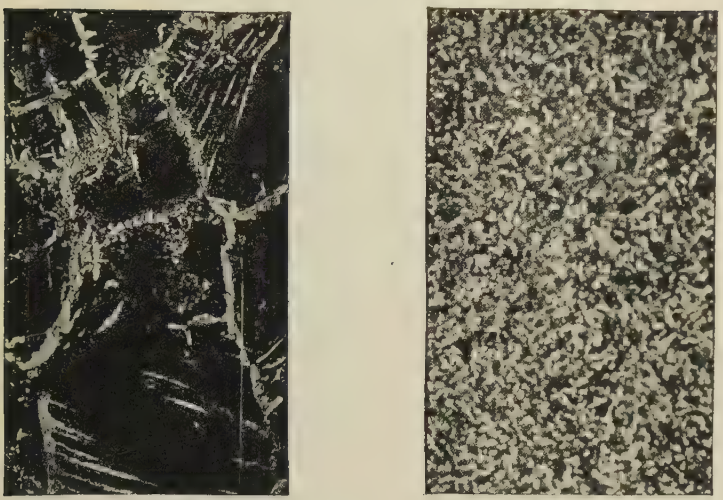
× 100 (Bullens).

Fig. 8.—Refining of overheated steel.

Refining Severely Overheated Steel. Another characteristic initial structure before annealing is shown in Fig. 9*a*. It is evident that this steel has been very severely overheated. Note the spine growth of the excess ferrite.

In *small* pieces, an unusually thorough soaking at the ordinary annealing temperature will entirely break up the old structure and refine the steel. This is shown by 9*b,* representing the result of annealing the 1-in. bar of Fig. 9*a,* in the manner indicated above.

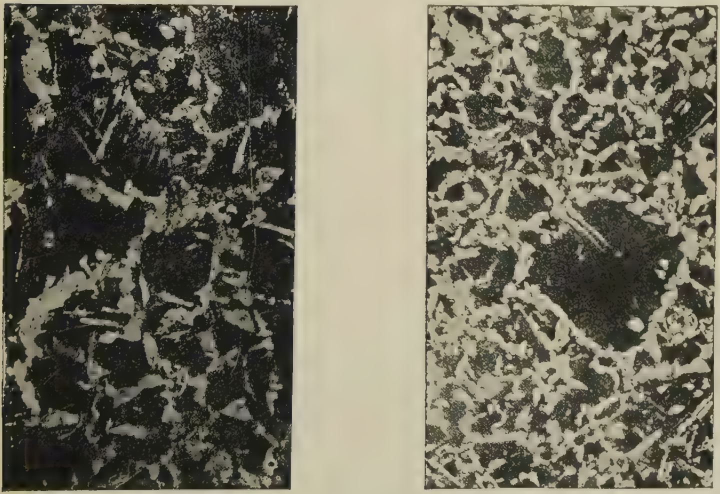
The annealing of large sections with a similar initial structure presents much more difficulty. The result of an ordinary double annealing of a 10-in. shaft, which had an initial structure similar to that



(*a*) As forged.

(*b*) Same, 1 in. bar refined by one long anneal. × 100 (Bullens).

FIG. 9.—Refining of a small section of severly overheated steel.



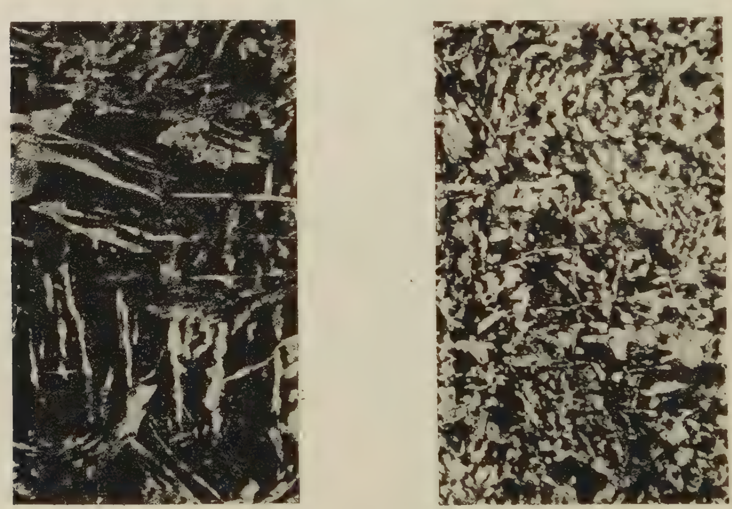
(*a*) After one double treatment, 1450° and 1400°.

(*b*) Same, after repetition of the double treatment. × 100 (Bullens).

FIG. 10.—Difficulty of refining a large section of overheated steel. This was a 10 in. diam. shaft, with initial structure like that of Fig. 9 (*a*).

of Fig. 9*a*, is shown in Fig. 10*a*. The grain size still is large, although the tensile results indicated good ductility. A second double anneal produced the results indicated in Fig. 10*b*. Although the steel largely is refined, there still is the tendency for large grains or "islands" here and there to remain unaffected.

Refinement of Maximum Overheated Steel. A still worse initial structure is represented by Fig. 11*a* and indicates that the steel in this 10-in. shaft is "burnt," or, at least, has been heated to the maximum temperature possible without being burnt. A structure like this, even in small pieces of steel, is very difficult to refine, whereas in sections of large dimensions it often is economically impracticable. Figure 11*b* shows the result obtained from a 10-in. shaft after eight annealings at various temperatures, totaling some 200 hr. of soaking. Even after such repeated heating and cooling, the original, distinctive arrangement of the ferrite has not been eliminated entirely!



(*a*)
Very badly overheated initial structure of 10-in. diam. shaft.

(*b*) × 100 (Bullens).
Same, not completely refined after *eight* anneals.

FIG. 11.—Another example of the difficulty of refining large sections of overheated steel.

Special Double Anneal. When the initial structure previous to annealing is of such character that a simple anneal, or a double anneal at normal temperature, will not refine the steel, raising the temperature of the first anneal often is found of practical benefit. In such a case the first heating is carried to a temperature considerably (200° or even 300°) in excess of the upper critical; after thorough soaking, the steel is cooled to a temperature below the lower critical, followed by a regular full anneal. The technique of the first anneal is based on the fact that the changes take place the more rapidly the higher the temperature. By this means the time may be shortened, for diffusion and equalization will take place more rapidly. Commercial success in breaking up initial overheated structures is dependent upon obtaining the maximum diffusion and equalization in as short a time as possible. Such high temperature treatment will produce a larger grain size than would a more normal temperature, but, if the overheated structure, by a subsequent operation, can thus be broken up, the mere refining of a structure of large grain size is comparatively easy.

Whether or not the steel shall be air cooled or furnace cooled after the first heating is a question which must be determined largely by experience. If the soaking at the high temperature has been sufficient entirely to obliterate the initial structure, then it is not so material which method of cooling is used. But, if complete diffusion and equalization has not taken place, it would be considered good practice to air cool the steel in order that as little opportunity as possible may be given for the old structural outlines to reappear. On the other hand, it might be argued that, with an initial structure similar to that of Fig. 11*a*, effort should be made to secure full separation of the excess ferrite into grains (as in Figs. 4*d* and 5*c*) and by such means attempt to eliminate the long, banded streaks of ferrite. The method and equipment for handling the stock in and out of the furnace, the value of the time required by furnace cooling, and the distinctive plant conditions as well as the size and shape of the work will have a strong bearing upon the method of cooling to be used.

A banded structure is sometimes present with streaks of ferrite in hypoeutectoid or of cementite in hypereutectoid steels appearing in the structure. (See Fig. 6*c*, Chap. 4.)

Such a structure looks too nonhomogeneous to be tolerated, though the properties of the banded steels may not be widely different from those of unbanded ones, and considerable study has been given to the homogenization of this structure. Long soaking at about 2200° for ferritic and at about 2000° for cementitic steels is necessary for complete homogenization. A discussion of the causes for banding is given by Epstein.[3]

The description in the preceding pages of the methods that may have to be resorted to when a recalcitrant large forging is to be annealed should not lead the reader to fear that annealing is ordinarily such a difficult task. In general, the annealing of pieces of ordinary section requires but a single treatment. When the heat treater does meet one of the uncommon, difficult cases, he may be at his wits' end to deal with it. It is the importance of these cases rather than the frequency of their occurrence that justifies the space given to their discussion.

Treatment for Strength-Ductility. Provided that the initial structure of the steel is such that an ordinary full anneal will produce a small grain size, the best combination of ductility, strength, and machining properties which can be obtained for C steels without an actual quenching operation will be obtained by the following operations of normalizing and drawing.

Heat to slightly over Ac_3, air cool to below Ar_1, and reheat at a temperature slightly below Ac_1 (about 1250°), holding the steel at this temperature until it is heated uniformly throughout, and then slow cool, or air cool, as there will be little or no change during cooling from a temperature under the lower critical.

By permitting the steel to air cool, after the first heating, to a temperature below the critical, advantage is taken of the retardation in the transformation of austenite into coalesced ferrite and pearlite. This effect will increase with the percentage of C, and will be greater the smaller the mass-surface factors of the piece, so that there will be the tendency to form a mass of emulsified pearlite and ferrite rather than the lamellar structure. The reheating to a temperature below the lower critical, if not prolonged unduly, will neither change the grain size nor allow of the coalescing of the excess ferrite or of the individual constituents of the pearlite. At the same time, as will be explained in Chap. 7 there will result an excellent combination of high ductility, good strength, and machining properties.

Annealing for Maximum Ductility. If, regardless of tensile strength, the maximum ductility is desired, and the ordinary full anneal will produce a small grain size, the method should be as follows:

Anneal at a temperature slightly above that of the Ac_3 range, slow cool to a temperature under that of Ar_1, reheat to a temperature slightly above that of Ac_1, and cool in the furnace for a spheroidizing anneal.

The term "spheroidizing" has been more commonly applied to hypereutectoid steels, but spheroidizing also produces maximum ductility in the lower C or forging grades, and the mechanism of the process is the same regardless of the C content.

In the 1917 edition of this book various annealing methods were described for obtaining maximum ductility in heavy forgings through the use of what actually might be termed "spheroidizing annealing," although then not called by that name. The recommended treatments consisted of a full anneal or normalizing process, according to the character of the initial structure, followed by a thorough soaking either just below, or in and slightly above, the Ac_1 range. Satisfactory spheroidization and quicker response occur when a temperature—1400°—slightly above Ac_1 is used. Results from this treatment on a 0.40% C steel have been given beneath Fig. 8*c*.

Spheroidizing Hypoeutectoid Steel Under Ac_1 Another method has been used in published work [4] on the spheroidizing of large forgings in the lower C ranges. Using a forging analyzing 0.45% C and 0.57% Mn, there is shown plainly the laminated pearlite and free ferrite obtained from an anneal at 1475° (Fig. 12), and the spheroidizing through two prolonged heatings (48 and 72 hrs., respectively) at 1250° (Figs. 13 and 14). The tensile test results are as follows:

SPHEROIDIZING 0.45% C STEEL FORGINGS UNDER Ac_1, RADIAL TESTS

Treatment	Tensile	Yield	Elong. %	R. A. %
1. 1475° for 10 hr. Furnace cooled.	84,750	40,250	19.5	28
2. 1250° for 48 hr. Furnace cooled.	72,250	35,000	24.5	34
3. 1250° for 72 hr. Furnace cooled.	69,500	33,250	30.0	52

Although in general the spheroidizing anneal is best carried out by first annealing above Ac_3, slow-cooling below Ar_1, reheating slightly above Ac_1, and again slow-cooling, nevertheless, in very large forgings in which the temperature lag between center and outside is very great, a single high temperature anneal followed by prolonged and repeated heating under Ac_1 may be more suitable. It produces greater structural uniformity, but not as great ductility, though the difference in ductility is not as great as indicated by the figures cited, since those for the first method are on longitudinal and for the second on radial specimens.

Effect of Rate of Cooling upon the Pearlite. The rate of cooling from the annealing temperature has a very great effect not only upon the network, grain structure, and excess ferrite, but also upon the characteristics of the pearlite. The rate of cooling through the lower critical, at which the transformation of the solid solution into pearlite

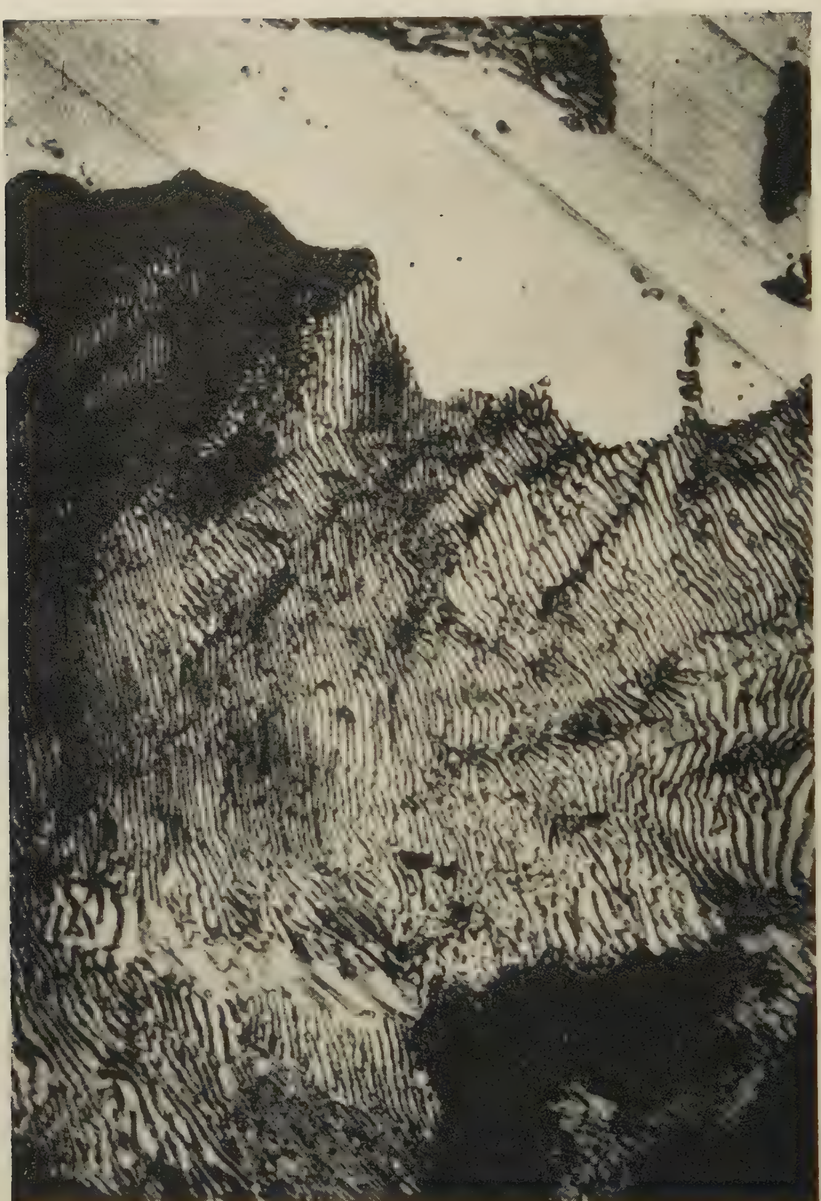


Fig. 12.—First Treatment. × 1000. (MacPherran and Harper.[4])

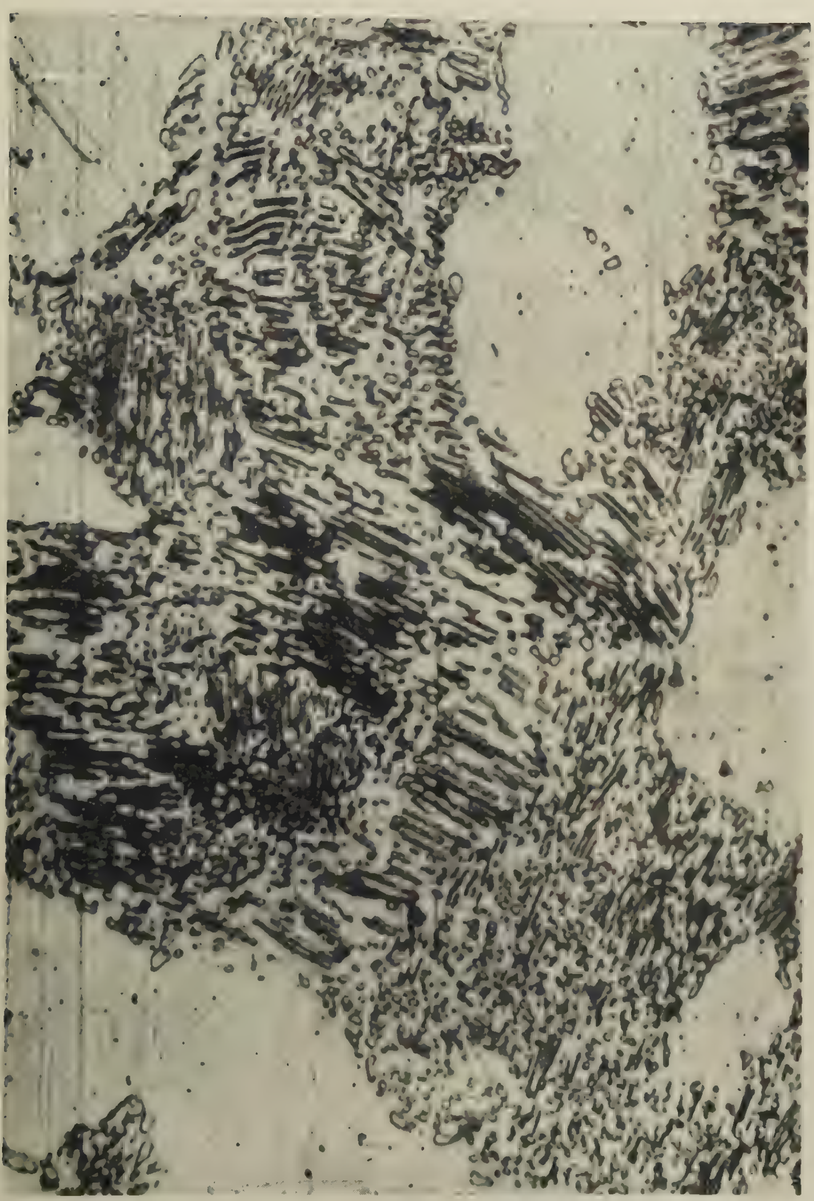


FIG. 13.—Second Treatment. × 1000. (MacPherran and Harper.[4])

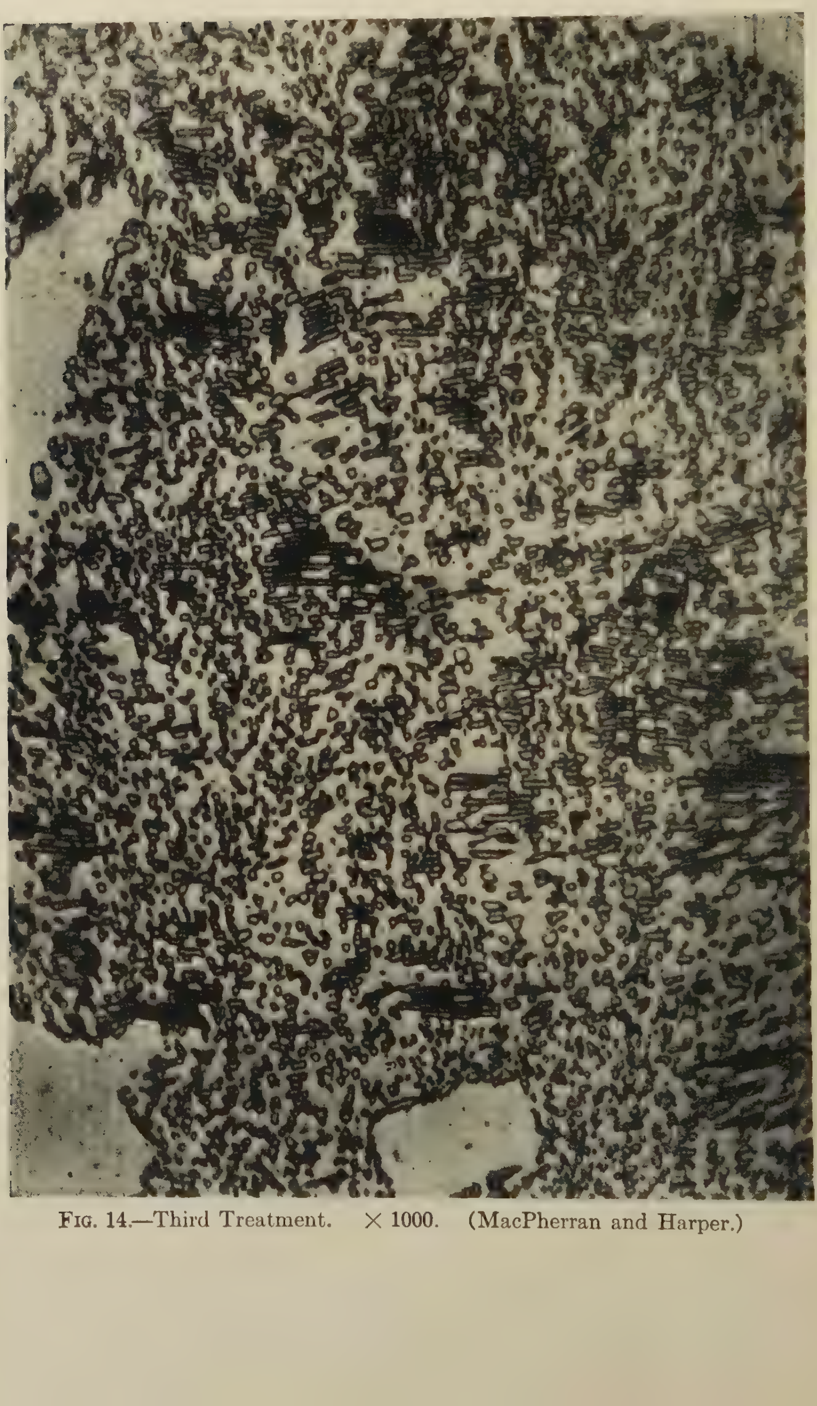
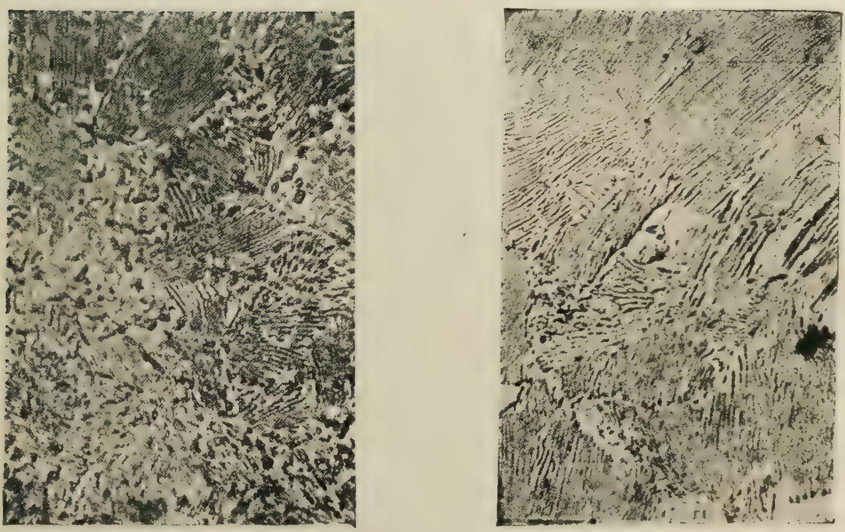
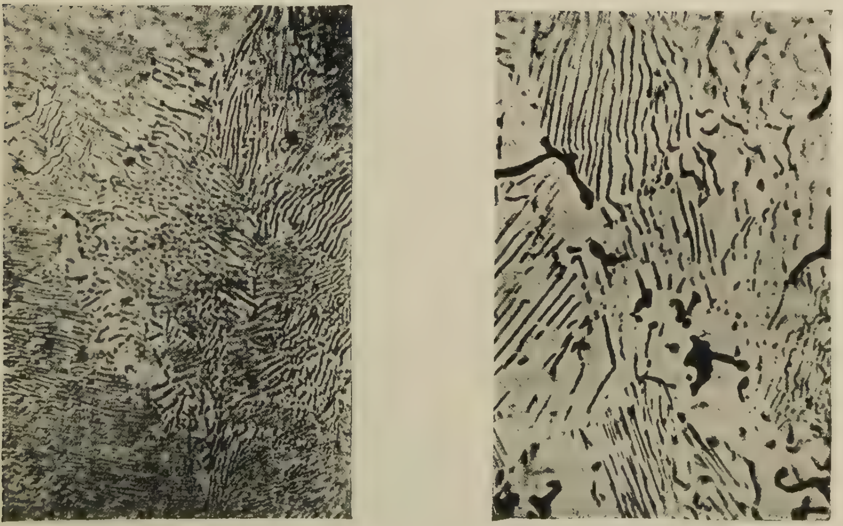


FIG. 14.—Third Treatment. × 1000. (MacPherran and Harper.)



(*a*) Rapid cooling. (*b*) Slower cooling.



(*c*) Still slower cooling. (*d*) Very slow cooling.

× 1000 (Van Tassel and Price)

Fig. 15.—Pearlite in 0.73% C steel, cooled at different rates from the annealing temperature.

is effected, will so change the thickness of the layers of ferrite and cementite in the pearlite that widely varying physical results may be obtained in this manner, as would be expected from the differences in structure shown in Fig. 15. The effect will be discussed in Chap. 5.

Annealing Semi-hard Steel. The annealing of hypoeutectoid tool steel—say, 0.55-0.75% C—may be broadly grouped under two headings, dependent upon the initial condition of the steel and upon the results desired. Such steel, if it has been carefully forged, is strengthened by the mechanical working; a full anneal—that is, heating at a temperature over the upper critical—will entirely obliterate the results of the forging operation. If, therefore, it is desired simply to anneal the steel in order to put it in suitable condition for general machine work—that is, to soften it and at the same time to retain the grain structure due to forging—the anneal should be carried out below the critical, or in the neighborhood of 1250° to 1300°, i.e., a process anneal is used. On the other hand, if it is desired to obtain the finest grain size possible and to eliminate entirely the effect of any previous heating or forging work, the anneal should be at a temperature slightly above the critical, or in the neighborhood of 1450° to 1400°, dependent upon the composition of the steel in question.

Protection of Steel. One of the vital points in obtaining a satisfactory steel after annealing is the protection of its surface. Steel, when heated beyond a low red heat, exhibits a great tendency to oxidize or scale, this action increasing in the presence of O_2, with the temperature and the length of time involved. This condition is exaggerated in open flame furnaces operated so as to produce sharp instead of soft, slightly hazy flames. Decarburization to a depth of ¼ to ½ in. is not a rare occurrence on very large sections, requiring very long soaking periods, where improper combustion and heat application are the rule. If, because of poor furnace design and worse operation, such conditions do exist, in order to produce a clean surface it is necessary to protect the exposed surface of the steel in some manner. Bar steel is often annealed in a tube carefully packed with charcoal and the ends closed with caps or luted with clay.

On the other hand, the prevention of *excessive* oxidation or scaling of the metal during the heating process is attainable with the *proper furnace design and operation*. Assuming such a design, if the furnace is operated so as to produce soft, hazy flames, there should be little occasion for packing the steel in charcoal or similar substances, unless the surface appearance is important and will warrant the extra cost for tubes and the extra time, fuel, and labor incident to their use.

A large proportion of such oxidation occurs during the cooling process. This is due, no doubt, to the air which reaches the steel after the outward pressure in the chamber has been released. In practice this condition has been partly offset by placing pieces of wood in the chamber, with the idea of creating a less oxidizing atmosphere around the steel while it is cooling.

Box Annealing. For the protection of larger masses or a number of smaller pieces, "box annealing" is often resorted to. This applies particularly to cases where a finished surface must not be injured. The steel is placed in a pot or box made of cast iron, of plates riveted together; or of cast, welded, or mechanically dovetailed construction, of heat-resistant alloys. This box may or may not be lined with some refractory substance such as silica brick. If warpage is feared the metal is packed carefully with some material such as ground mica, sand, charcoal, charred bone or leather, lime, etc. If the steel is low in C a carbonaceous substance is usually avoided, for a slight case-hardening action might take place. In the higher C steels, and especially tool steels, carbonaceous materials may be used, although it is better to mix the charcoal with clean ashes. Sand and ground mica are probably the most satisfactory of the simple refractory materials. The cover then is placed on the box and the box with its contents is charged into the furnace and given the proper degree and duration of heating. The box should be raised from the floor of the furnace so that the hot gases may have opportunity for circulation around it. When annealing is complete, the box may be removed from the furnace and allowed to cool to atmospheric temperature.

The protection of the surface against scaling or decarburization has been given much attention and "controlled atmospheres" have come into use to avoid these damaging effects of air and furnace gases. Controlled atmospheres will be discussed in Chap. 12.

Stead's Brittleness. It was previously stated that practically no change occurs below the Ac_1 range if no previous hardening of the steel has taken place. The one exception is that of very low-C steels and is due to the fact that steels very low in C behave more like pure or carbonless iron, there being but small percentages of cementite (and therefore pearlite) to influence the grain size.

If these low-C steels are given a certain small amount of deformation by cold work and are then heated in the annealing range just below Ac_1, the ferrite grains *are* coarsened. The grain growth is a function of time and temperature. Smaller or larger deformations, outside the zone of "critical strain," do not produce this growth.

Such critically strained low-C steels, held at, say 1100°-1200° for a considerable length of time, will develop such coarsening of grain size as to make the steel unfit for commercial use if any degree of strength is required. This phenomenon is known as "Stead's Brittleness."

Samuels[5] has described cases in which excessive grain growth occurs in low-C steel without the steel having been deformed, but when it has been cooled at a certain rate during normalizing prior to box annealing. With steels of higher C content practically no change occurs until the Ac_1 range is reached.

Summary of Annealing Methods. The following charts give a condensed summary of the methods of annealing previously discussed; they should be used only in conjunction with a careful consideration of all the elements involved and should be altered to suit conditions.

Reference Key to Annealing Methods

Temperature

Key	
A	Ac_3 max. + 50°, or Ac_3 end + 15°
B	Ac_3 + 200° to 300°
C	Ac_1 + 25°
D	Ac_1 − 50°

(Steels inherently resistant to grain growth may be annealed at higher temperatures.)

Soaking Time

N	Normal
P	Prolonged

Cooling

Q	Air
R	Retarded

Initial Structure

	Character	Illustration Reference	Page
I.	Reasonable	Fig. 5*a*	107
II.	Overheated	Fig. 8*a*	110
III.	Severely overheated	Fig. 9*a*	111
IV.	Maximum overheated	Fig. 11*a*	112
V.	Cold-worked	Fig. 6*a*	108
VI.	Castings	Fig. 11*a*, Chap. 2	91

X. Repeat treatment indicated if results of previous treatment are not satisfactory.

Annealing Methods

Steel	Results Desired	Small Sections	Large Sections
I.	Commercial full annealing (maximum refinement and high ductility).........	*ANR*	*ANQ*
II.	Commercial full annealing (maximum refinement and high ductility).......	*APR*	*ANQ·ANQ* or *ANR·ANR*
III.	Commercial full annealing (maximum refinement and high ductility).......	*APR* or *ANQ·ANR*	*BNQ·ANQ* *XANQ*
IV.	Commercial full annealing (maximum refinement and high ductility).......	*BNQ·ANR*	*BPQ·APQ* *XAPQ* or *XBNQ·ANQ*
V.	Commercial full annealing (maximum refinement and high ductility).........	*APR*	*APQ*
VI.	Commercial full annealing (maximum refinement and high ductility).........	*ANR*	*APR*
I.	Good machining only..................	*DNQ*	*DNQ*
I.	Maximum ductility and refinement......	*ANR·CNR*	*ANR·CNR*
I.	Strength ductility......................	*ANQ·DNQ*	*ANQ·DNQ*
VI.	Commercial annealing.................	*ANR*	*APR*

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CHAPTER 4

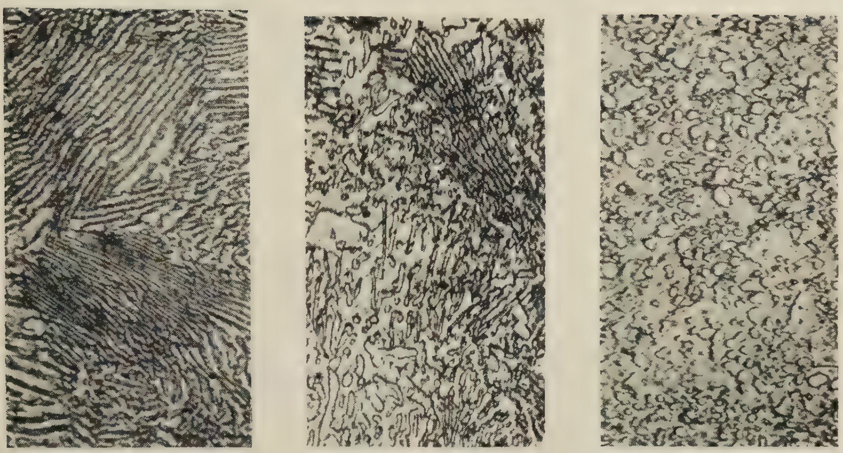
SPHEROIDIZING: TOOL-STEEL ANNEALING

Spheroidizing. Spheroidizing is a process by which, under suitable conditions of temperature and time, the cementite of pearlite "balls up," or coalesces, into globules, or "spheroids." Spheroidizing materially affects the ductility of the lower C steels. This will be discussed in further detail later in the chapter. In the realm of the higher C steels, spheroidizing is of vast importance.

Effect on Machinability. The annealing of C tool steel—or, more specifically, of eutectoid and hypereutectoid steel—differs markedly from the annealing of the constructional or hypoeutectoid steels, in that it more definitely affects the cost of subsequent machining operations. The annealing process as applied to the lower C steels is connected generally with strength requirements, either directly in relation to the annealing operation itself, or as a preliminary to further heat-treatment operations, likewise having in view the meeting of strength requirements. The annealing of tool steel (say, 0.75% C and upward), however, not only involves the grain refinement necessary to avoid cracking and distortion in the drastic hardening operation and to provide maximum strength in the finished tool, but also—and generally of equal or greater importance—to provide easy machinability. These factors are dependent directly upon the pearlite modifications, varying from the more common plate or laminated effect, through the intermediary stages of the balling-up of the cementite, to a final stable stage of complete spheroidization. This is illustrated by Fig. 1*a*,*b*,*c*, and the following table:

Characteristic Annealed Eutectoid Steel

Figure	Structure	Brinell Hardness	Machinability
1(*a*)	Laminated pearlite	228	Poor
1(*b*)	Mixture of laminated pearlite and spheroidized cementite	196	Fair
1(*c*)	Cementite completely spheroidized in ferrite ground mass	163	Good



(*a*) Laminated pearlite. Annealed above Ac_1.

(*b*) Mixture of laminated pearlite and spheroidized cementite.

(*c*) Spheroidized cementite with ferrite ground mass. Final anneal just below Ac_1.

FIG. 1.—Annealed eutectoid steel. × 500.

The economic value of spheroidizing therefore is evident. It subsequently aids materially in obtaining a hardened product of uniform quality.

Spheroidizing Methods. From the practical standpoint, the mechanics of spheroidization are correlated directly with initial structure in its relation to the lower critical, Ac_1, and with time as an important function.

Spheroidal cementite may be obtained either from the laminated pearlite stage or from the emulsified stage, as follows:

I. From Laminated Pearlite:

(*A*) The reheating temperature should be a little below or very slightly above Ac_1. In this case, the process is comparatively slow inasmuch as the time required is considerable for the cementite plates to divide and each subdivision to spheroidize. Study of Fig. 1*b* will indicate the mechanics of the process, showing the rounded particles formed and the process of forming from the plates of cementite. This spheroidization can be accomplished below Ac_1, but the time required is long.

(*B*) If this maximum temperature exceeds a definite limit above Ac_1, the cementite will appear again as laminated pearlite;

(*C*) If the temperature does not reach Ac_1, spheroidizing will not readily take place.

II. From Emulsified Pearlite:

In a very finely divided or practically emulsified * mixture of ferrite and cementite, such as in sorbite (produced by quenching and tempering), or in emulsified pearlite, the nuclei from which the spheroidizing process can start are already established. The cementite plates are so tiny that they easily form such nuclei. For this reason a temperature may be used:

(*A*) Just under Ac_1 say 1250° to 1300°, with certainty; or
(*B*) At or slightly above Ac_1 as noted in (I), using a shorter period time, but with a narrower margin of safety.

In both (I) and (II) the degree of spheroidization is a function of time.

These differences in the spheroidization of lamellar pearlite and of non-lamellar structures account for the better stability (i.e., better creep-resistant properties) of lamellar pearlitic steels for high temperature service, not withstanding the better room-temperature strength of the latter. Normalized and drawn steels usually show better creep properties than steels of the same composition that are quenched and tempered. Years of service at temperatures well below those that will produce rapid spheroidization may finally produce it.

Effect of Initial Structure. With the above essential principles in mind, it is possible to determine upon the method of procedure from the character of the initial structure. The metallurgist's problem too often is one of trying to repair the damage done in the steel mill through improper heating and cooling in relation to the work at hand. (See discussion in Chap. 3 and Chap. 14.) Dependent, therefore, upon the preceding history of the steel—such as temperature of heating for cogging, forging, or rolling; the amount of such mechanical work; the finishing temperature; and the rate of cooling from the finishing temperature—the "initial structure" may be coarse or fine and with or without a network of free cementite or free pearlite. If it is coarse, the pearlite may range from massive to somewhat emulsified; if fine, the emulsified pearlite may be coarse to fine, and, in fact, may even be on the way to spheroidization. Fine structure is usually associated with the emulsified type, inasmuch as finishing in or near the critical range usually produces such structures in the common-sized, tool-steel bars, owing to the fairly rapid rate of cooling through the Ar range.

Tool-steel Annealing. In Fig. 2*a* the coarse grain size with massive, laminated pearlite and a cementite network is shown. A normal anneal at 1350° partly spheroidized the steel (Fig. 2*b*) but the time was not of sufficient duration to equalize the structure and complete the spheroidization. Better results would have been obtained either by (1) a longer heating at the same temperature, (2) a double anneal, or (3) a normalizing (air cooling) from a higher temperature—to produce an emulsified structure—followed by a normal spheroidizing anneal.

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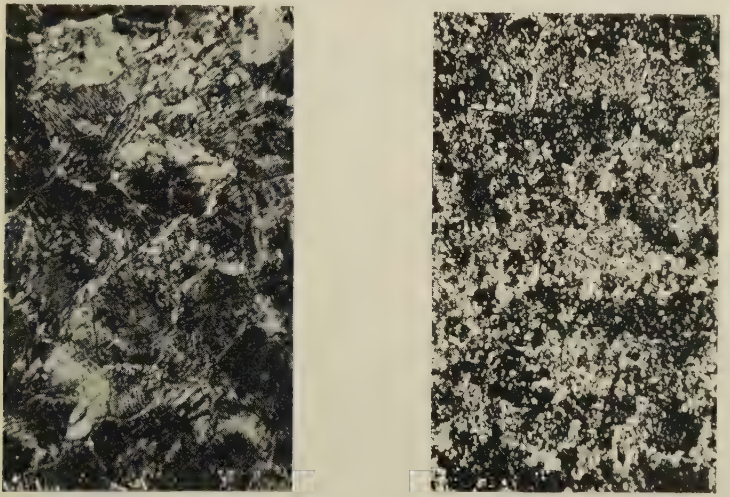
* This will be defined in Chapter 5.

The emulsified structure of Fig. 3*a*, although coarse grained, as indicated by the network outline, will respond to the 1350° spheroidizing anneal more readily than that of Fig. 2*a*. Although the final structure (Fig. 3*b*) is not as fine as might be desired, because of the initially coarse grain size, the Brinell hardness of 174 would be satisfactory for general machine work.



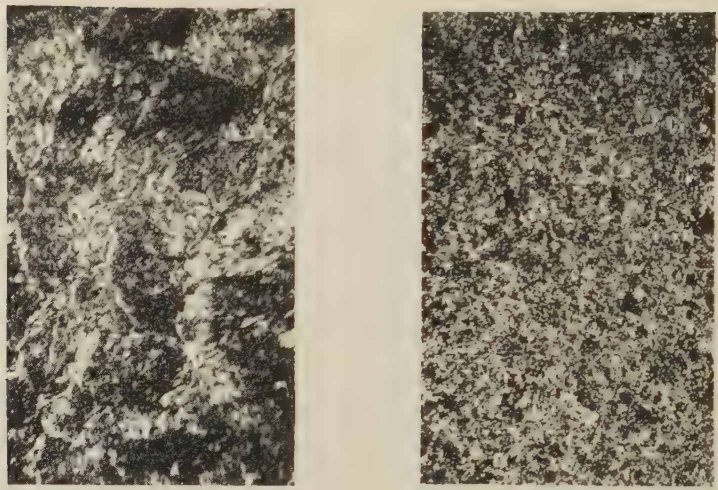
(*a*) As rolled, coarse grain, massive, laminated pearlite. ×500

(*b*) Same, partly spheroidized at 1350°. Not uniform. ×150

Fig. 2.—Incomplete spheroidizing anneal of a 1.10% C steel.

Figure 4*a* illustrates a 1.10% C steel initially emulsified though coarse, and without network outline indicative of large grain size. Annealed at 1350° for a comparatively short time, Fig. 4*b*, it immediately started to spheroidize, giving a Brinell of 174; but the transition was not completed. A somewhat longer heating would have completed the spheroidization.

The finely emulsified structure of Fig. 5*a* is ideal for spheroidizing, if, indeed, the process may not be considered as already commenced. Thorough soaking at a temperature under Ac_1 is sufficient

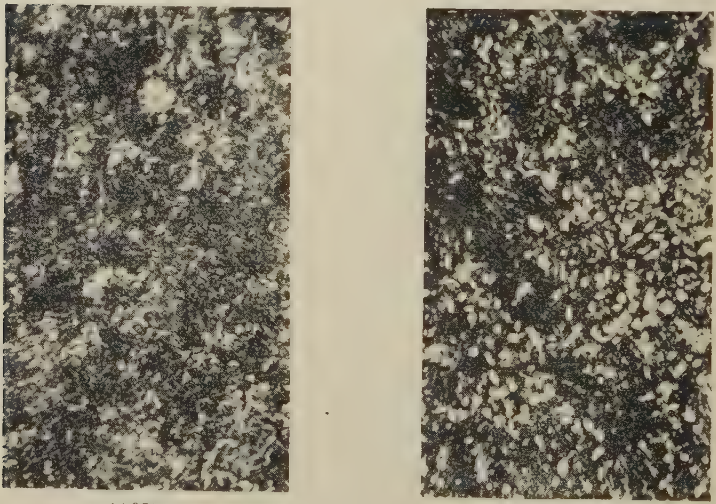


(a) Initial structure coarse, but partly emulsified.

×500

(b) Same, annealed at 1350°. (174 Brinell.)

FIG. 3.—Annealing of 1% C steel of good initial structure.



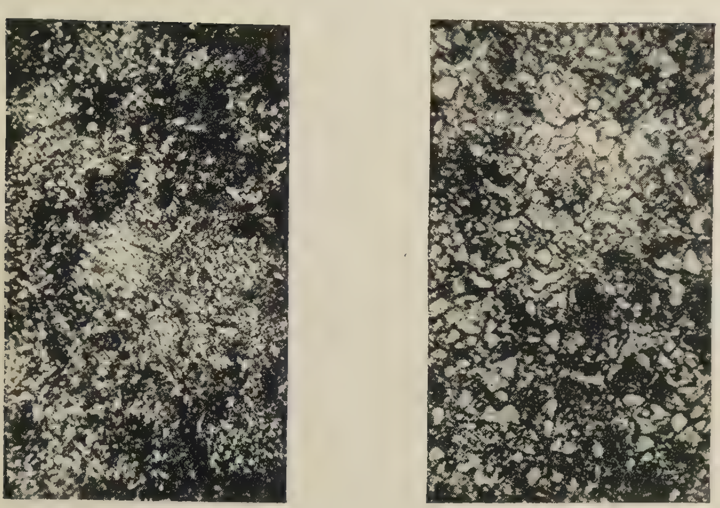
×650

(a) Initial structure, as rolled, well emulsified.

×520

(b) Same after very short anneal at 1350°, showing good response. Anneal not completed.

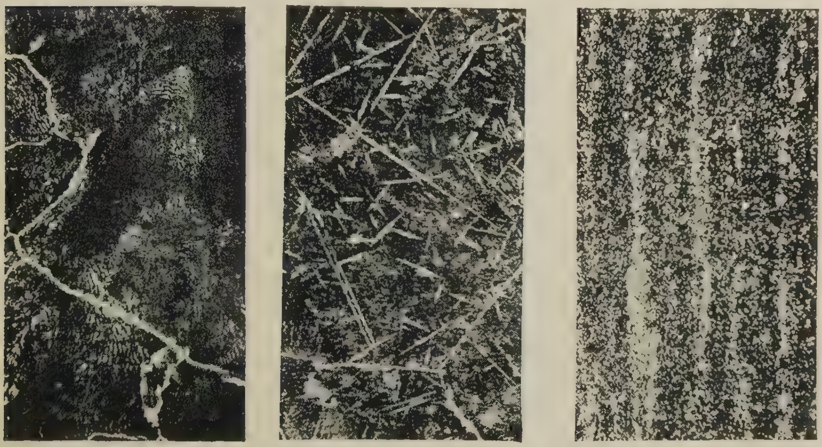
FIG. 4.—Another case of good initial structure. 1.10% C steel.



×520
(*a*) Initial structure, as rolled, fine, well emulsified.

×650
(*b*) Same, spheroidized at 1250°.

Fig. 5.—Spheroidizing of 1% C steel of very good initial structure.



(*a*) Excess cementite in network.

(*b*) Excess cementite in spines.

(*c*) (Green.) Excess cementite in lines, due to cold finishing.

Fig. 6.—Examples of poor initial structure in high C tool steel.

for complete spheroidizing. This is illustrated by Fig. 5*b*, obtained by spheroidizing at 1250°.

Spheroidizing High-carbon Steel. In Chap. 3 it was shown that free or excess ferrite in the hypoeutectoid steels could only be taken completely into solution at a temperature above the upper critical, Ac_3. Similarly, in the hypereutectoid steels, the free or excess cementite is taken completely into solution only when the temperature exceeds its upper critical, known as Acm. This solubility line is shown in the chart, Fig. 7, Chap. 2. It therefore is apparent that normalizing from above Acm is necessary as a preliminary to spheroidizing in hypereutectoid steels, especially of the higher carbons, under the following circumstances:

1. When they have been finished at excessively high temperatures and the excess cementite has precipitated almost entirely to the network (Fig. 6*a*);
2. When they have not had sufficient working to eliminate from the ingot the cementite spine effect (Fig. 6*b*); or
3. When they have been finished so cold as to produce cementite lines (Fig. 6*c*) similar to the ferrite lines of a cold-worked soft steel.

Therefore, after the steel has been heated above Acm and the excess cementite dissolved and then retained (emulsified) by either air or oil quenching, the spheroidizing treatment may be undertaken as usual, generally just under the Ac_1.

Another effective method of spheroidizing hypereutectoid steels is to heat and soak above Ac_1, but below Acm, to dissolve the cementite of the pearlite without dissolving the excess cementite that is beyond solubility limit at that temperature in the austenite. This excess cementite will tend to spheroidize. The steel is then air cooled or quenched and again heated, this time to just below Ac_1 for final spheroidization of the cementite that had been dissolved in austenite.

CHAPTER 5

NORMALIZING AND "AUSTEMPERING"

Normalizing. In the discussion of annealing it has been pointed out that in order to secure ready absorption of excess ferrite and equalization of the C in austenite, it is necessary or advisable in the cases of steel castings, of overheated forgings, and of very large forgings, to heat to much higher temperatures than in regular annealing, even though in so doing the austenite grain size may become very large. To avoid the coarse structure that would result from the slow cooling of regular annealing, more rapid cooling, in the open air, is employed. Similarly, in C tool steels and in case carburized steels a preliminary heating to a high temperature, above Acm, may be advisable to break up the continuity of the cementite network surrounding the pearlite, and air cooling is used since slow cooling would reestablish the network.

The same procedure of air cooling may be used even upon steels that would readily homogenize at ordinary annealing temperatures, when a finer structure is desired than that resulting from slower cooling. When the more rapid cooling is to be used, so that a fine structure will result anyway, it is not as important to avoid the grain growth of austenite that comes from higher heating, but it is desirable to attain homogenization of the austenite. So homogenization is secured by heating to a higher temperature for a shorter time.

The rate of cooling of austenite that has been given a good degree of homogenization profoundly affects the structure. Confining ourselves in this chapter to such rates of cooling that austenite all decomposes to ferrite and cementite while passing through Ar_1 and none is retained to change into martensite; (i.e., there is no quench hardening), we find that changes in the cooling rate will alter the way the ferrite and cementite are deposited as they form from the austenite. Thus different structures and different properties may be obtained.

Cooling Rates. Cooling rates are generally vaguely described as "fast" or "slow," or by mentioning the way they are produced, e.g., by furnace cooling, cooling in ashes, in still air, in an air blast, quenching in oil, in water, or in a high-pressure water spray. These phrases have to be further clarified by stating the mass of the steel being cooled.

A very fine wire will lose heat faster in ashes than the center of a huge forging would under the most drastic water quenching. The really important matter is the *rate* at which the steel cools through the critical range, and this can be expressed in degrees per minute. Unfortunately, the measurement of cooling rates is not easily carried out; so the qualitative phraseology ordinarily has to be employed. We have some idea of the speed of travel when we use the words "at a walk," "double-quick time," automobile driving speed, aircraft cruising speed, speed of a bullet, although each of these varies, and we know the true speed only when it is definitely expressed, as in miles per hour.

A few determinations of cooling rates are on record. Howe, Foley, and Winlock[1] found that very slow furnace cooling is at the rate of about 2° per min. and, in air cooling of small pieces in an air blast, the cooling rate is around 120° per min. To harden a 0.2% C steel fully, as by quenching, requires that it cool faster than about 2400° per sec., just as it passes through the critical.

Mechanism of Separation. In a hypoeutectoid steel, as austenite cools below Ar_3 it starts to deposit its excess ferrite, and this proceeds till all excess ferrite has been rejected around patches of austenite that have finally become enriched in C to the eutectoid concentration. If the cooling from Ar_3 to Ar_1 is slow, the ferrite has time to grow into a connected or coalesced network with islands of 0.85% C austenite inside. If the cooling is fast, the patches do not so readily grow together and remain peppered more uniformly throughout the mass. At Ar_1 the C-bearing austenite has to break up into ferrite and cementite.

If the cooling at Ar_1 and just below it is extremely slow, as in a spheroidizing anneal, the cementite tends to contract, by something analogous to surface tension, into balls within a matrix of connected ferrite. (See Fig. 1*c*, Chap 4.) In any but this excessively slow cooling the austenite splits up into alternate layers of ferrite and cementite, the lamellar structure of Figs. 1, Chap. 2; 14, Chap. 3, and 1*a*, Chap. 4. But the thickness of each layer is less, the more rapid the cooling. Hence in furnace cooling we get coarsely lamellar pearlite, easily resolved into the layers at rather low magnifications. With more rapid cooling the plates become thinner and thinner, more of them packed into the same space, but the plates are still quite flat. Patches of pearlite have the eutectoid ratio, about 0.85% C, on very slow cooling; but rapid cooling gives pearlite of other ratios. It is possible to have pearlite (in hypoeutectoid steels) with as low as, say, 0.60% C. In such cases the estimate of the C from the amount of pearlite shown on a micrograph is not accurate. Still more rapid cooling gives not only still thinner plates, but warped ones, so that their cross section in a micrograph looks curly. Still faster cooling brings the warpage and the thinness of the plates to such a degree that, instead of layers, the microscope shows a confused, emulsified mass. This emulsified stage has been called, in the older nomenclature (used in previous editions of this volume, in common with all books of their time), a "sorbitic structure." A still further stage, in which dark-etching clumps appear, has been called "primary troostite," * or "troostite of quenching," since it appears when quench-hardening rates of cooling are nearly approached.

The Term "Sorbite" Reserved for Quenched and Tempered Structures. Modern microscopic study has shown these latter stages to be nothing more or less than the ferrite and cementite of pearlite, in nonlamellar form. Vilella, Guellich, and Bain [2] suggest that the terms sorbite and troostite should not be used for the structures resulting directly from decomposition of austenite, but should be reserved for the decomposition structures of martensite on tempering.

Under this nomenclature, after austenite in cooling becomes of eutectoid or near-eutectoid composition, it decomposes to pearlite (unless it is retained to go on down to martensite, or even to be retained as austenite itself, under the influence of the rapid cooling rates of quenching). We shall follow Vilella, Guellich, and Bain in limiting the term "sorbite" to a structure born from martensite, but shall retain the terminology of the troostites.

The stages then range, with increased cooling rates, from coarse pearlite (coarse in respect to thickness of lamellae), through fine pearlite, on to curly pearlite, and finally to fully emulsified pearlite.

When the cooling is fast enough to produce the curly or the fully emulsified pearlite, it is also fast enough to avoid coalescence of ferrite into a network, and the ferrite is jumbled up with the pearlite.

Properties of Pearlite. The continued subdivision makes increasingly for strength, and whereas ductility may be a trifle lower than with coarse pearlite and coalesced ferrite, it is high.

The strength varies with the width of the ferrite and cementite plates in the pearlite, as Belaiew [16] and Green [17] have shown. If they are about 0.4 micron † wide, in a eutectoid steel, it shows 100,000 tensile. If they are 0.25 micron wide, the strength rises to 150,000. As the width of the plates decreases, the warping of the plates into the curly structure goes on, finally reaching the highly emulsified state of *primary troostite* with a tensile of some 225,000.

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* The terms troostite, sorbite, martensite, and austenite are all named in honor of pioneer metallographers, Troost, Sorby, Martens, and Austen.

† A micron is 1/1000 of a millimeter, i.e., 0.000039 in.

Hence the emulsified pearlite which is obtained by air cooling of small specimens of C steel, or whatever approach to it we can get in larger specimens, is valuable as a structural material. The heat treatment to produce it is short; hence normalizing, i.e., heating above Ac_3 and air cooling, is much employed. It is used as a prelude to full annealing or to quenching, in order to put the steel into such state that the ferrite does not have to look very far away for the C that is coming into it from the pearlite on a reheating to form homogeneous austenite preparatory to the final cooling, be that cooling slow or rapid. It is used also by itself, without any further re-creation of austenite, for hypoeutectoid steels. For hypereutectoid steels it is generally used only as a prelude.

Normalizing is applied to the thinnest and the thickest pieces of hypoeutectoid steel we have to heat treat.

Normalizing of Sheet Steel. Very low C, unalloyed steel is used for tin plate, automobile body sheet, and the like. Such steel will run around 0.10% C, or below, too low for effective hardening by ordinary quenching means. Such steel must be very ductile, since ductility is required in forming seams on tin cans or in stamping an automobile top or fender. Strength is usually of little relative importance, but the forming properties are all-important.

In modern practice the thin sheets are hot rolled to perhaps $\frac{1}{16}$ in. singly, or in continuous mills as strip, then either pack rolled hot, or cold rolled continuously to final gage. Even the pack-rolled material has some rolling strains and needs to be annealed, whereas the cold-rolled sheet must be softened in some way.

Just as in the heavy forgings discussed in the previous chapter, the maximum of ductility and formability is obtained by a full anneal; but, since grain growth and resultant roughness of the formed surface ("orange peel") come from a full anneal, one is restricted to annealing below the critical, or to normalizing. Severely cold-rolled strip, because of its strained and deformed structure, softens on annealing at 1200°, whereas pack-rolled material requires higher temperatures, sometimes up to 1400°. In either case in the regular annealing cycle, stacks of sheets or coils of strip, often in lots of 10 to 50 tons, are covered with a box or bell into which a neutral or non-oxidizing gas atmosphere is led (as will be more fully discussed in Chap. 12) and slowly heated to the annealing temperature by electric resistors, by radiant tubes, or by direct firing outside the box. Very slow heating is required to avoid overheating of the edges because of the poor thermal conductivity of the piles or coils, and, for the same reason, they cool very slowly. The heating and soaking cycle is usually some 20 to 40 hr., and it takes another 24 hr. or more to cool to a low enough temperature, about 280°, so that the charge will not oxidize when the box is opened. Keelan[3] has discussed the schedules for strip annealing in detail.

In contrast to this, single sheets or continuous strip, can be raised to a normalizing temperature of 1750° in 1½ min, held 1 or 2 min., cooled to 800°, reheated to 1150° for 1 or 2 min., and then cooled, all in a space of less than 15 min. Or, a cycle can be used of 1½ min., heating to 1750°, holding 1 min., cooling in 1½ min. to 1020°, retarding the cooling so that it takes 2 min. to drop to 840°, then cooling to 280° in 3 min.[4]

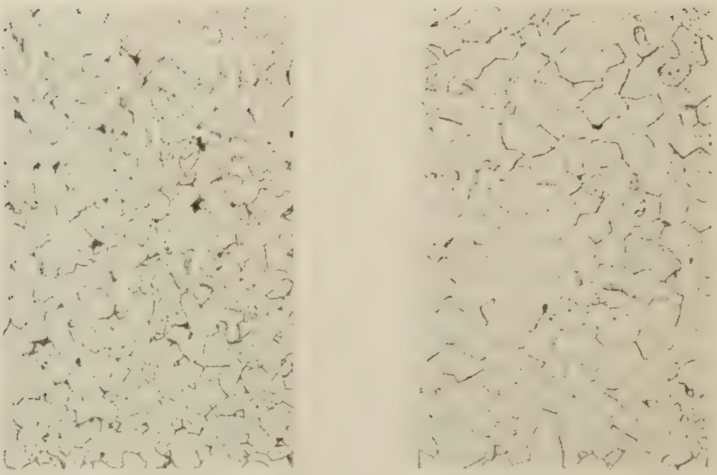
The reheating to 1150° or the retarded cooling from 1020 to 840° in this case is probably not employed for quite the same reason that steels of higher C content are reheated or drawn after normalizing but more probably in order to control precipitation-hardening effects. (See Chap. 8.) Ferrite has a very small solubility for cementite, so that at temperatures above the solubility line *OP* of Fig. 7, Chap. 2, some of the cementite that came out at the grain boundaries on the first cooling can be redissolved on reheating above this line but below Ar_1 and by controlled cooling can be precipitated within the grains instead. Other elements, such as O_2 and N_2, present in very small amounts in these unkilled, rimming steels, have rather similar solubility curves in ferrite to that shown in Fig. 7, Chap. 2, and the treatment will probably redissolve them all.

Such normalizing gives formability almost equivalent to that of box annealed materials, with fine grain and hence good surface on the stampings. The same steel, normalized and box annealed, is shown in Figs. 1*a* and *b*.

Normalizing of Steel Castings. Although full annealing, i.e., slow cooling but from a high temperature, has been the usual treatment for C steel castings, normalizing is now more in vogue. The air cooling of fairly heavy castings, particularly when a whole carfull is pulled out of the oven in which they were heated, is rather slow, but yet enough more rapid than a furnace cooling to give different properties.

Sims[5] cites an example of two structures obtained from the same heat of cast medium Mn-V steel, shown in Fig. 2*a*-*b*. Note the uniformity in tensile properties as contrasted with the improvement in impact with the finer grained steel.

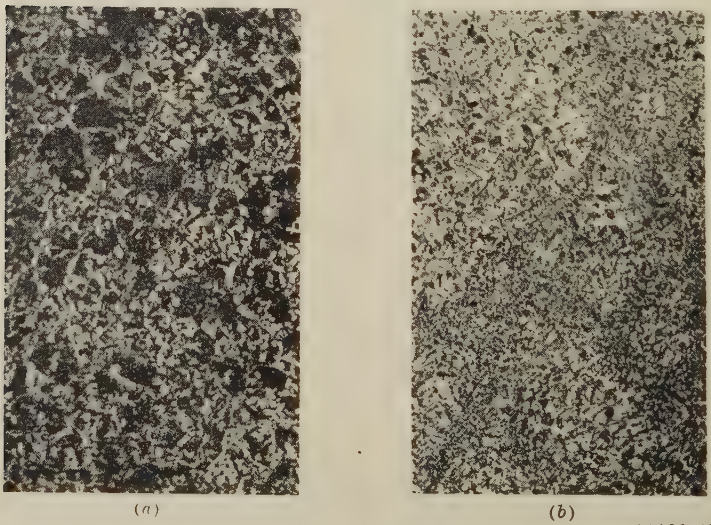
The importance of sufficient heating prior to normalizing to homogenize the steel is brought out by Fig. 3*a*, *b*, *c*, *d*, also from Sims. The



(*a*) Normalized, 1850°, 15 min., air.

(*b*) Box annealed 1450°; heated and cooled at 50°/hr.

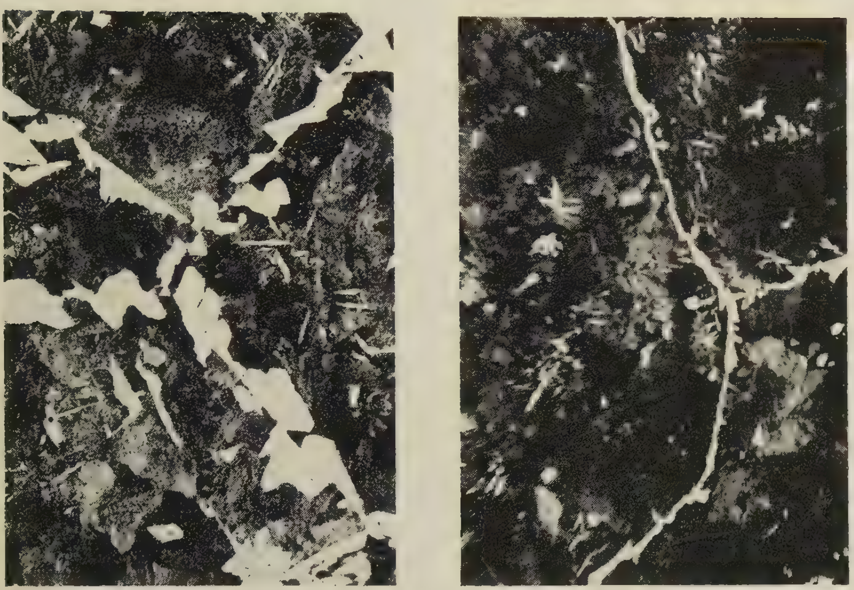
FIG. 1.—Structure of 0.07% C steel. ×100 (Battelle).



×100 (Sims 5)

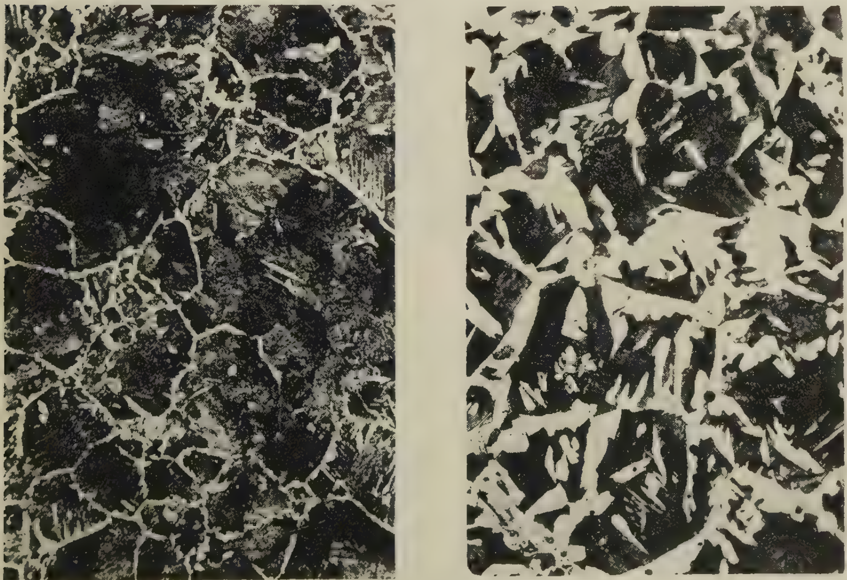
Tensile	98,000	101,500
Yield	63,000	66,000
Elong. %	27	27.5
R. A. %	55.5	57.5
Izod, ft.-lb.	18.5	48

FIG. 2.—Structure vs. impact resistance of two specimens from the same heat of Mn–V steel. Both specimens were double normalized, but the second normalizing on (*a*) was from an excessively high temperature.



(*a*) Center of 1-in. cube, as cast. ×100

(*b*) Center of 6-in. cube, as cast. ×25



(*c*) Same as (*a*) after 1650°, 1 hr., air. ×100

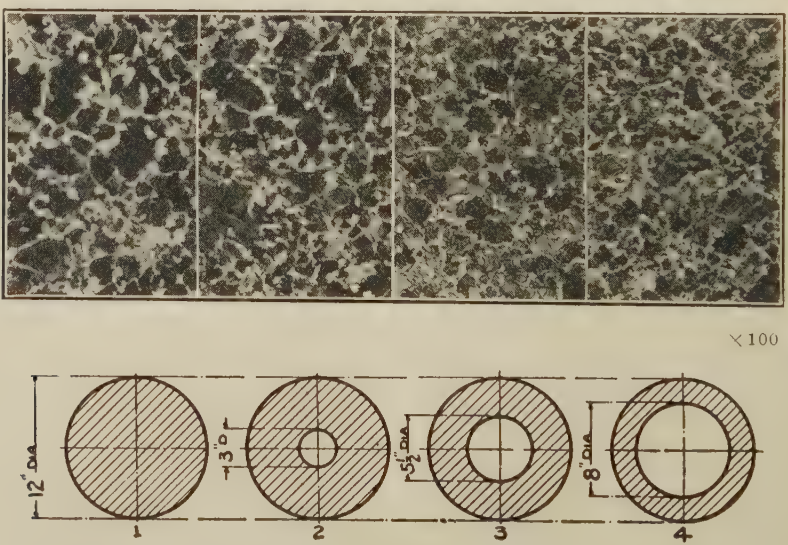
(*d*) Same as (*b*) after 1650°, 1 hr., air. ×100

Fig. 3.—Normalizing of light and heavy sections of the same heat of cast steel. (Sims.[5])

raw structure in a cast 1-in. cube was somewhat refined after 1 hr. at 1650°, but it took 6 hr. at 1650° to obliterate the raw structure in the center of a cast 6-in. cube. The grains in the latter are still coarse and a second normalizing would be required to produce a structure like that of the 1-in. cube.

Normalizing of Forgings. In heavy forgings, such as locomotive driving axles and the like, one would like to develop the maximum combination of properties, but quenching to secure it can hardly be used on C steels, for the objects are so massive that C steel will not harden throughout. Moreover, unless the piece is symmetrical, it is likely to crack on quenching. In forgings for big guns alloy steels are employed that will harden much more deeply on quenching, so that the guns may be quenched and tempered. When the cost of the steel and of the treatment is such a factor as to eliminate this, and C or moderately alloyed steels that would not harden throughout are used, these huge forgings are either full annealed as discussed in Chap. 3 or they may merely be normalized.

Fry[6] has discussed locomotive forgings of 0.34% C steel in a manner that clarifies the question of structure and properties versus rate



	1	2	3	4
Ratio volume to surface	3:1	2.25:1	1.6:1	1:1
Tensile	96,500	97,000	97,500	97,000
Yield	55,000	56,000	59,000	60,000
Elong. %	21	23	25.5	25
R.A. %	34	36	42	40.5

Steel of 0.34 C, 0.29 Si, 0.58 Mn, 0.034 P, 0.04% S.

FIG. 4.—Effect of boring on mechanical properties and structure of normalized forgings of the same steel. (Fry.[6])

of cooling in normalizing. Forgings 12 in. in diameter by 3 ft. long were left solid, or bored out as shown in Fig. 4, heated together to 1500°, held there 6 hr., and air cooled. The more rapid cooling as the section is decreased improves the structure and the ductility.

Tempering after Normalizing. The normalizing treatment is usually followed by a tempering, below the critical, to remove cooling stresses.

No apparent structural change occurs on such tempering, but more or less marked improvement in ductility results. Data which show this, kindly supplied by the International Nickel Company, are tabulated below:

Forged Steel—0.30 C, 1.20 Mn, 1.60% Ni

	Normalized from 1550°	Normalized 1550° and Tempered 1000°
Tensile	94,000	94,000
Yield	73,000	71,000
Elong. %	27	29.5
R. A. %	51.5	58

Cast Steel—0.29 C, 0.74 Mn, 3.57% Ni

	Annealed	Normalized 1650°	Normalized 1650° Tempered 1200°	Double Normalized 1650° Tempered 1200°
Tensile	103,000	111,800	103,000	102,250
Yield	65,750	80,750	71,000	74,500
Elong. %	22.5	12.5	22	24
R. A. %	41.5	15	46	47

Cast Steel—0.35 C, 0.60 Mn, 3.55% Ni

	Annealed	Normalized 1650°	Normalized 1650° Tempered 1200°	Double Normalized 1650° Tempered 1200°
Tensile	104,000	114,700	104,500	105,000
Yield	64,500	72,500	71,500	73,250
Elong. %	22	12	22.5	23.5
R. A. %	42	19	45	50

Although this set of data does not show the changes in impact, they are likely to be even more marked than the changes in static ductility. Cast steels with less of a tendency toward air hardening than the Ni-Mn steels are not so notably improved by tempering after normalizing.

Stress Relief of Normalized Steel. The elimination of stress is of particular importance in the case of rotors for steam turbines. The forgings may be as large as 5 ft. in diameter by 20 ft. long and, since, for efficiency, the turbine is designed with very small clearances, any appreciable warpage or growth would cause rubbing of the moving parts. The permanence of dimensions must be maintained over the life of the turbine, 20 years or more. The turbines operate at high temperatures, with superheated steam under high pressure.

Stress relief, like other heat-treatment processes, is a matter of both time and temperature, so the operating conditions offer opportunity for gradual warpage to relieve any stress that is left in the forging. The choice of the steel, and its treatment, must be calculated to minimize the retention of stress.

Mochel[7] points out that the stress introduced in normalizing will depend upon the temperature of the Ar range and upon the stiffness or plasticity of the steel at that temperature. Comparing a 0.42% C steel with the transformation at about 1250° with a 0.25 C, 2.70 Ni, 0.30 Mo, 0.10% V steel in which the transformation is not completed till the temperature has fallen to about 815°, and recalling (see Chaps. 9 and 19, Vol. II) that Mo increases creep resistance at elevated temperatures, i.e., the steel is stiffer, he concludes that the stresses in normalized forgings of the latter are much the greater and demand very long stress-relieving treatments. Even the C steel, with its greater plasticity, requires long treatment as is evidenced by the following program for treatment of 0.42% C rotor forgings of 40-in. diameter and larger.

> The forging is not allowed to cool, but immediately after forging is transferred to a hot furnace and held at 1600° for 30 hr., then air cooled to black. It is then reheated to 1260° and held there for another 30 hr., followed by furnace cooling.
>
> Rough machining is then done, and the forging placed in a cold furnace, slowly heated to 1480°, held 27 hr., again air cooled to black, reheated to 1200°, held 30 hr., and furnace cooled. This reheating is done with the forging suspended with its axis vertical in a tall furnace.
>
> Finish machining is then done and a final stress-relieving draw given, again with the forging vertical, at 1050°, holding 48 hr. at temperature and furnace cooling. That is, the rotors are double normalized and given three stress-relieving draws.

Normalizing vs. Annealing. Although the ductility shown in a tensile test may be greater with full annealing than with normalizing, the higher yield and tensile strengths accompanying the more rapidly cooled structure give a greater area under the stress-strain curve of the tensile test, that is, greater ability to absorb work. Coupled with this is the specific effect of the finer structure, the combination leading to greater toughness in the impact test. Howe and co-workers [1, 8] demonstrated this nicely.

Influence of Variations in Slow Rates of Cooling on the Mechanical Properties of Carbon Steels

Carbon, %	Cooling Rate, Deg. F. Per Sec.	Tensile	Yield	Elong. %	R. A. %	Charpy Impact, Ft.-lb.	Brinell Hardness
0.34	0.02	70,000	35,000	28	50	6	130
	2	78,000	45,000	34	60	22	150
0.52	0.02	85,000	42,000	27	38	4	160
	2	95,000	53,000	28	42	8	190
0.75	0.02	90,000	40,000	20	35	1.8	180
	2	120,000	55,000	16	28	2	230

As the table shows, the beneficial effect of the normalizing rate of cooling as compared with annealing is more evident on forging grades than on steels of higher C content.

Stability of Structure. Normalizing and then tempering below the critical but well above the temperature of service is a favorite treatment for carbon and low-alloy steels for use at high temperatures where great stability of structure is desired. The creep resistance of steel with lamellar pearlite is much above that of the same steel with its cementite spheroidized. Bailey [9] has emphasized this and has shown that spheroidization may occur well below the critical temperature if the steel is held at temperature many thousands of hours. Thus a creep test of the steel for a thousand hours or so, in which the steel still retains its original structure, leads to the expectation of a certain order of creep resistance, but if on very long heating the structure shifts toward the spheroidized condition, the creep resistance falls. The stability of the original structure is an important thing to know in evaluating steel for high-temperature service. The change should not be unexpected, since, like any other structural change on tempering, spheroidization is a time-temperature function. In ordinary heat treatment the time that can be given is at best relatively short, so that the obvious variable is temperature. In long-time service the time factor has a chance to show itself.

Although the theoretical explanation is not obvious, it is found that a high-temperature draw below the critical, but well above the temperature of service, tends to stabilize the structure against spheroidization at the lower temperature. A normalize and draw, although producing lower room-temperature strength than a quench and draw, usually gives much better high-temperature stability and continued, effective, creep resistance.

Tempering after air cooling is, of course, applied also to steels that air harden, i.e., those whose critical cooling rates for formation of martensite are so lowered by alloying that an air cool is actually a quench. Air hardening is not properly called normalizing, because it does not put the steel in the "normal" pearlitic state. Air hardening will be discussed in a later chapter.

Effect of Cooling Rate after Welding. Even though air hardening, in the sense of the production of martensite, may not occur, C steel of more than 0.15% C, and still more notably, low-alloy steels with more than that amount of C, may be hardened, i.e., very fine pearlite produced, when the zone next to a weld cools from the welding heat. As will be discussed in Chap. 3, Vol. II, when a welded specimen shows any zone over 200 Brinell, a stress-relieving draw is used, in best practice, to prevent having a brittle joint. Large welded pipes are stress relieved by annealing at around 1200°. Special induction heaters have been designed, hinged to allow their being placed around the completed weld, so that the joint may be annealed in place after welding.

When steel is cut by a torch, the metal adjacent to the part that has been melted out will cool rapidly and become hardened if the C or alloy content of the steel is sufficiently high. Flame-cutting outfits are often supplied with a lower-temperature, diffuse flame which follows the cutting flame and makes the metal adjacent to the cut cool more slowly.

Patenting. A rate of cooling designed to give pearlitic structure is applied to wire to be further fabricated by cold drawing. The wire, usually of rather high C, from 0.60 to 0.85% C, is heated to a high temperature and a certain degree of grain growth is intentionally produced. Temperatures may be as high as 1850° or even 1925°, but the cooling is at such a rate as will give emulsified pearlite (i.e., what in earlier parlance was called a sorbitic structure). This rate of cooling can be obtained in air on the usual sizes of wire, but a steel with a slightly greater propensity toward hardening, i.e., one with Mn raised above the usual range, is usually chosen when air cooling is to be applied. A quench in hot lead held at around 900 to 1150° provides the proper rate of cooling and a more controllable one than air cooling. The structure obtained is then broken down by the wire drawing, the original grains drawing out into a fibered structure in the finished wire. The properties of lead-patented, cold-drawn wire are slightly superior to those of air-patented, cold-drawn wire. Either the air- or the lead-cooled wire is better for this particular purpose (of being cold-drawn) than wire of much smaller grain size produced by quenching to martensite and tempering back to sorbite. Discussion of patenting is given by Epstein[10] and by Sisco.[11]

"Austempering." A controlled rate of cooling involving quenching to and holding at a selected temperature, at which austenite is transformed into emulsified pearlite, is utilized in what has been termed "austempering." This is related to patenting in that the steel, heated into the austenite range, is plunged into a bath of molten lead or salt, held at a temperature sufficiently below the critical so that the steel is cooled through the critical range fast enough to prevent it from transforming to lamellar pearlite, but the bath is held at a temperature definitely above that at which martensite is formed. With a sufficient holding time, the austenite transforms to emulsified pearlite.

Because of the relatively elevated temperature, no martensite forms; hence the minute cracks that may be found in martensite, presumably due to the volume change occurring as it forms, are absent from the steel. It is thought that these cracks are not fully healed in quenched steel on tempering the martensite to secondary troostite or sorbite. Transformation of austenite to emulsified pearlite without passing through martensite, hence without these tiny cracks, thus produces a steel of greater toughness for a given hardness than do the operations of quenching and tempering. Compare Fig. 14, Chap. 9.

The cooling rate which makes "austempering" possible is a rather crucial one, and a suitable range of cooling rates extends inward from the steel surface only something like ¼ in. in C steels. Hence C steels will respond completely to this treatment only in thicknesses of about ½ in. or less. Somewhat greater thicknesses can be handled in alloy steels, but this subject has not yet been thoroughly explored. The time of holding at temperature must be sufficient to allow the austenite to transform so completely that practically none will be left to change, on further cooling, to martensite. This may require from a few minutes to a few hours in C steels, depending on the composition and the hardness sought, which fix the temperature to be used. Alloy steels may be so sluggish that the necessary time becomes impractical. The treatment is most effective in high-C steels and is chiefly applied to those of 0.60 to 1.20% C, since these are the ones in which quenched martensite is most likely to show the damaging minute cracks.*

Temperatures for Austempering. The steel is forced to transform at a selected constant temperature. The temperature for formation of primary troostite, without martensite, varies with the C content; e.g., to produce a Rockwell C hardness of 45, a 0.50% C steel will be held at 650° and a 1.10% C steel at around 1000°. Steels of 0.70% C and up can can be given a hardness of 50 Rockwell C, i.e., about 500 Brinell, by holding at 600 to 700°. The remarkable improvement in toughness so produced is shown below.

STEEL 0.74 C; 0.37 Mn; 0.14 Si; 0.04 S AND 0.045% P

Treatment of 0.18-in. Diam. Stock

	1450° into Lead at 580°	1450° into Oil Drawn ½ Hr. 600°
Rockwell C	50½	50
Tensile	282,500	246,500
Yield	151,000	122,000
Elong. % in 6 in.	1.9	0.3
R. A. %	34.5	0.7
Impact *	35	3

* Ft.-lb. absorbed, 0.18 in. round, un-notched bar.

In the bend test, the quenched and tempered rod snapped after being bent through only a small angle whereas the austempered rod took a right-angle bend without fracture. Davenport [12] shows several striking photographs of bent specimens.

It will be noted that the most notable improvement in the ordinary tensile properties is in the reduction of area and that the bend test reflects this, as was pointed out in Chap. 1.

The difference in toughness, between the austempered and the quenched and tempered conditions is most marked at and below Rockwell C 50; at Rockwell C 55, the un-notched round bars showed 5 and 1 ft.-lb., respectively for the two processes.

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* Hughes and Dowdell,[13] using steels of 0.34 C, 0.44% Mn and 0.49 C, 0.78% Mn, in 0.505-in. test-bar section, did not develop, in these lower carbons and larger section, properties superior to those secured by ordinary quenching and tempering.

Davenport[12] finds also that the endurance limit is improved; a 0.93% C steel was treated in 0.2-in. diameter bar to 50–51 Rockwell C showing 108,000 lb./in.^{2} endurance limit when this hardness is obtained by quenching and tempering, and 136,000 when obtained by "austempering." By comparing the tensile strength to be expected at this hardness with the endurance limit, it will be seen that the "austempered" material has not shown the drop in endurance ratio usual above 400 Brinell but maintains the normal ratio. The 136,000 lb./in.^{2} endurance limit is about the highest ever reported.* (See Fig. 12, Chap. 1.)

Because of the limitations as to thickness of piece and composition of steel that can be effectively treated by this method, it is not of wide applicability but has important specific applications. Interesting shovels have been made by this process, for example.

Information on this process is available in publications by Davenport[14, 15] and co-workers. The principle upon which it is based will be brought out in the discussion of the "S" curve, Chap. 6. (See also Chap. 16, Vol. II.)

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* Later data (E. E. Legge, The Industrial Application of Austempering, *Metals & Alloys*, v. 10, 1939, pp. 228-242) do not substantiate the higher endurance of austempered stock as a general thing.

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CHAPTER 6

PRINCIPLES OF HARDENING

Elements of Hardening. In order to discuss intelligently the mechanism of hardening, it is necessary to obtain a clear understanding of the reactions taking place during both the heating and the cooling of definite steels through the critical ranges. The heating phase is a preparation for the cooling phase, and, without the existence of proper preparation in the former, the results from the latter will be useless. The most that the cooling phase can do is to retain and develop the characteristics imparted by the heating phase.

An effort will therefore be made (1) to illustrate, in an elementary manner, by chart and micrographs, the elimination of coarse grain size, network, free ferrite (in hypoeutectoid steels), and associated characteristics of untreated steel, together with the necessity for complete diffusion and equalization of the solid solution; (2) to show the relationship between hardening and annealing; (3) to explain the mechanism by which the solid solution passes through the critical range on cooling and the austenitic transition products or "hard constituents" result; and (4) to discuss the factors affecting these results.

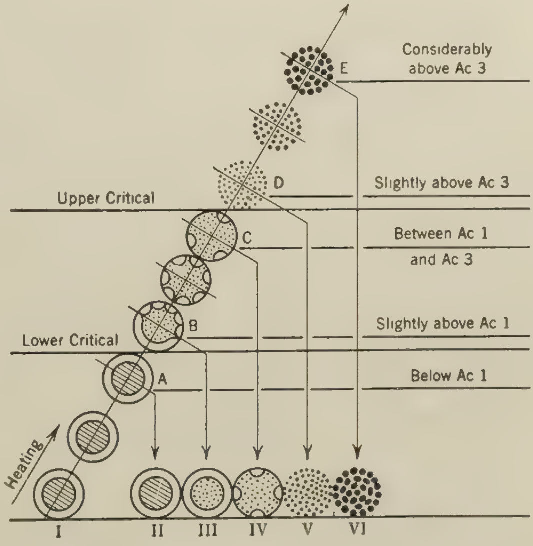

Fig. 1.—Graphical illustration of changes in a 0.40% C steel on heating and quenching.

Changes on Heating. In the chart, Fig. 1, Column I (at the left), represents a normal, 0.40% C, pearlitic steel (at the bottom of the column) and the structural changes taking place in that steel as it is heated progressively through and beyond the critical ranges. It is assumed that the structure and micrographic constituents obtained by heating to various temperatures, such as *A* to *E*, may be retained by rapid cooling, as illustrated by Columns II to VI.

Thus, heating to a temperature *A*, under that of the lower critical will produce no change * in the original steel, which consists of pearlite (the inner cross-hatched circles) and ferrite (the white area), with the outer circumference representing the grain size. The quenching likewise will produce no change, as is illustrated by Column II.

Heating to a temperature *B*, slightly above the lower critical, will change the pearlite to the solid solution (represented by the dotted area), but without affecting the free ferrite. Quenching, Column III, will therefore produce a semi-hardened steel—since the solid solution is the "hard" constituent—with a refinement only of the "ground mass" (the original pearlite).

Heating to a temperature *C*, between the lower and upper criticals, will effect a progressive absorption by the solid solution of the remaining free ferrite. Quenching, Column IV, will therefore produce a "harder" steel than in case III, but nevertheless without complete refinement of the steel as a whole. This is indicated in the chart by the solid line of the circle, representing the initial grain size, and by the islands of free ferrite along the boundaries.

Heating to a temperature *D*, slightly above the upper critical, if prolonged for a length of time sufficient to effect complete diffusion and equalization, will entirely transform the steel, giving it the smallest austenite grain size possible. Quenching, Column V, will retain this condition and give the maximum hardness possible. The absence of the definite circumferential line indicates graphically the elimination of the initial grain size and the production of many fine grains within the area originally represented as a grain.

Heating to a temperature *E*, considerably above the upper critical, will tend to increase the grain size; and quenching, Column VI, will give a more brittle steel.

Structures Produced. These reactions are illustrated by the group of micrographs given in Fig. 2. The original structure, consisting of pearlite (dark) and free ferrite (white), with a large grain size is shown at the right of Fig. 2*a*, which is taken at the passage of the steel through the lower critical: at the right the original structure remains unaffected when the specimen (heated at one end) is quenched from a temperature below Ac_1; at the middle a distinct change in structure

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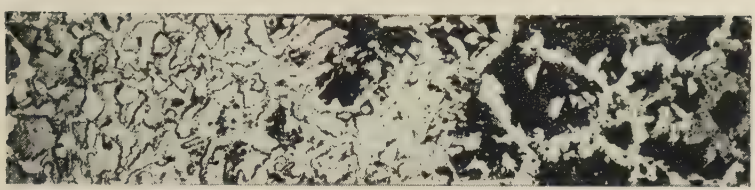
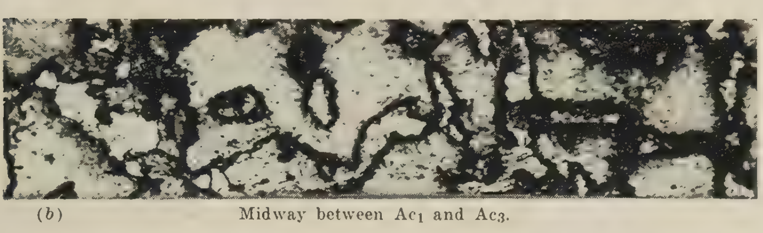
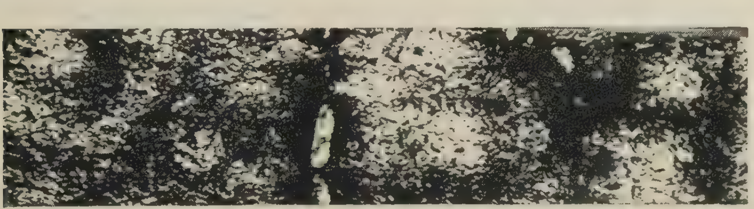
* Spheroidization can be produced below Ac_1 but the time required is long.



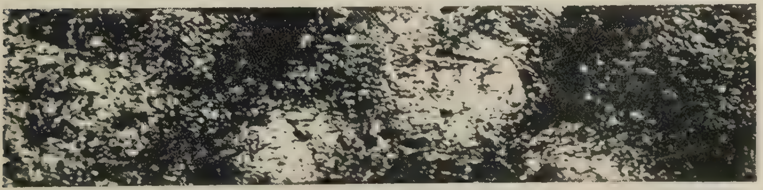
(*a*) Above Ac$_1$. Ac$_1$. Below Ac$_1$.



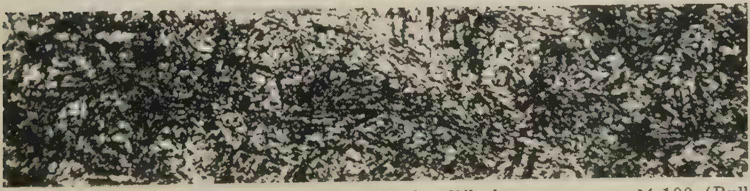
(*b*) Midway between Ac$_1$ and Ac$_3$.



(*c*) Ac$_3$ almost completed.



(*d*) Just above Ac$_3$, but lacking time for diffusion.



(*e*) Above Ac$_3$, with time for diffusion. × 100 (Bullens.)

Fig. 2.—Structures of the same steel, quenched over a range of temperatures.

shows clearly the Ac_1 transformation, the former dark pearlitic areas having given way to the solid solution, and the excess free ferrite (also white) remaining unaffected; further to the left the few remaining islands of pearlite are dissolved completely, thus marking the completion of Ac_1.

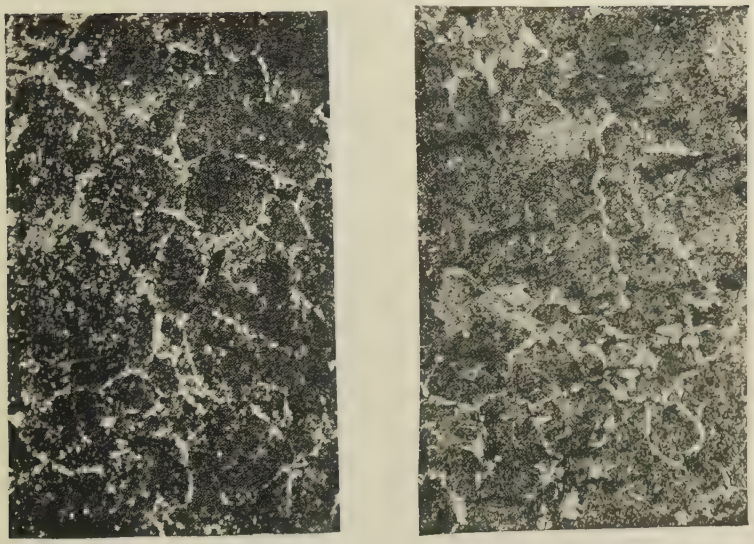
Figure 2*b* illustrates the structure about midway between Ac_1 and Ac_3. The amount of free ferrite is noticeably diminshed as Ac_3 is approached. Figure 2*c* is taken just as the last islands of unabsorbed ferrite are about to disappear at a temperature only a trifle below Ac_3. The black outlines show the location of the former free ferrite and illustrate the point that, although the ferrite has been absorbed into the solid solution, there has not beeen sufficient time for its diffusion.

Diffusion and Homogenization. The steel in Fig. 2*d* is taken just above Ac_3, the total absorption of the ferrite having taken place. The phenomena of diffusion and homogenization or equalization, however, are also brought out. Although the Fe has been dissolved by the solid solution, the solution is not yet homogeneous, as is shown by the difference in etching. If the steel at this particular location were to be tempered or toughened by reheating, these lines of high-Fe content would at once serve as centers of re-precipitation for the excess ferrite; a large grain size would be reborn; and the desired refining effect from heating for hardening would not be obtained.

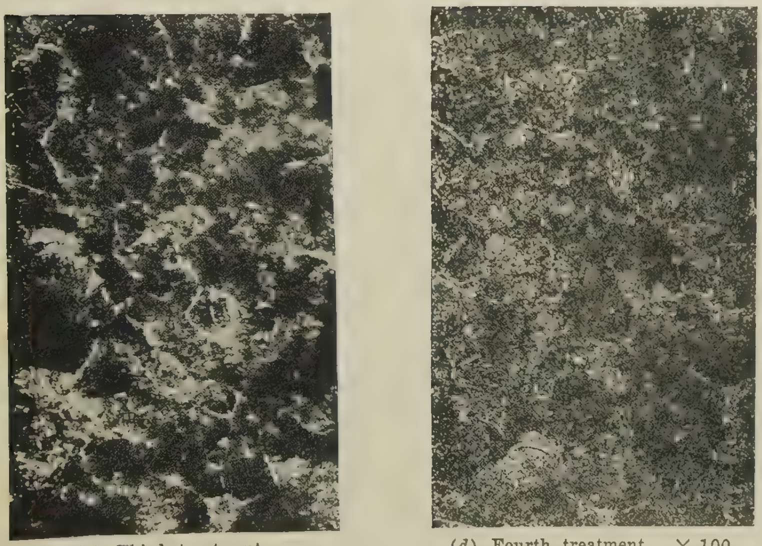
It is therefore evident that there must be a period of saturation sufficient for diffusion and homogenization to take place, so that the full effect of the Ac_3 end temperature can be developed in the steel and retained by subsequent quenching. Such treatment gives the structure illustrated by the Fig. 2*e*, in which uniform martensite is evident.

The practical influence of initial structure in relation to the factors of diffusion and homogenization is brought out by the micrographs of Fig. 3. This series, representing the effect of successive quenchings (and toughening) of a 10-in. diameter shaft, 0.40% C, illustrates how difficult it is to absorb and diffuse the ferrite in an initially poor structure when the initial structure, like that in Fig. 10*a*, Chap. 3, shows coarse grain size and overheating by the forge shop. The difficulty of annealing this shaft was referred to in Chap. 3.

Relation of Hardening to Annealing. During the *heating* operation the same changes take place in microscopic constituents and in the structure as a whole, without regard to whether the steel is to be softened thereafter by slow cooling or hardened by quenching. In the latter case the heating is likely to be to a higher temperature, but the main difference in the final results of the two processes is due to the

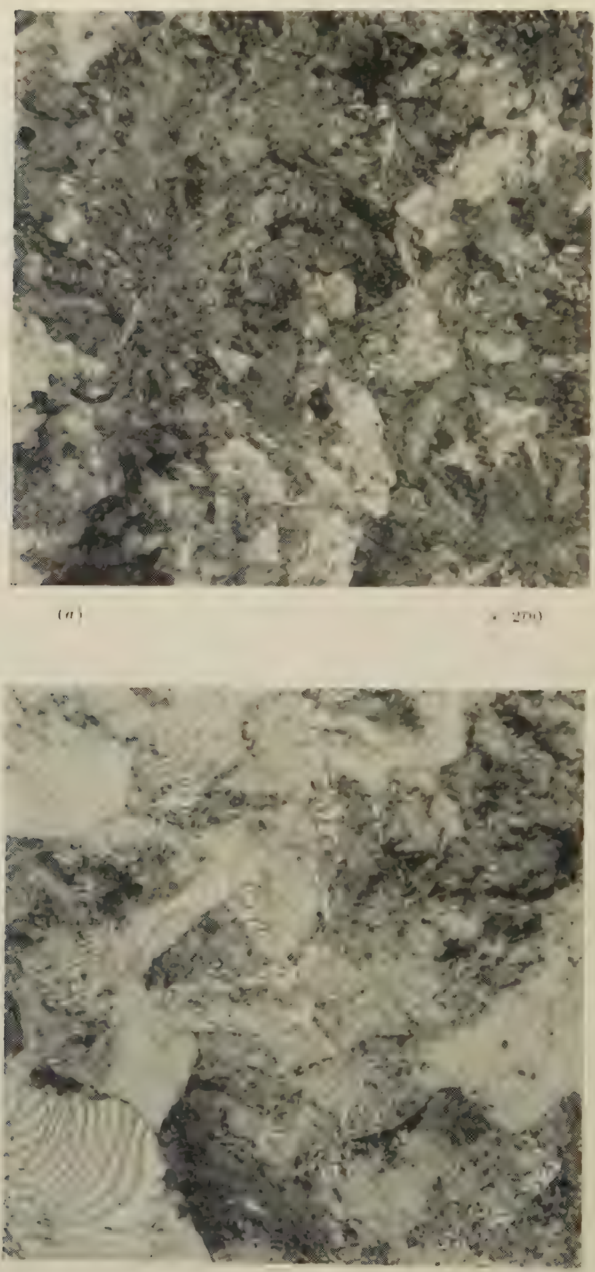


(*a*) First treatment. (*b*) Second treatment.



(*c*) Third treatment. (*d*) Fourth treatment. × 100. (Bullens.)

FIG. 3.—Successive re-treatments of a 10-in. diam. 0.40% C shaft of poor initial structure. See Fig. 10*a*, Chap. 3, for original structure.



(a) × 250

(b) × 1000

FIG. 4.—Primary troostite at two magnifications. (Lucas.[1])

rate of cooling through the critical, which alters the nature of the microconstituents retained in the steel when cold.

The effect of slow cooling through the critical ranges, which is characteristic of true annealing, has been discussed; in brief, the austenite or solid solution shifts its C content through generating proeutectoid ferrite (or cementite) to the eutectoid ratio of 0.85% C, and is then transformed with increase of volume at Ar_1 into pearlite, with which the ejected ferrite (or cementite) remains mixed.

With rapid cooling, however, and in combination with other factors later to be mentioned, the completion of the change of the parent solid solution (austenite, solution of C or carbide in gamma iron) will tend to be retarded and certain other products obtained. These are known as "primary troostite" and "martensite"; and it is these products which cause the steel to be "hard."

Therefore, the crux of the hardening process is to *prevent,* by suitable methods, prior breakdown of the austenite and the separation of pearlite therefrom, and to force it to transform instead into hard martensite.

Full hardening requires the formation of an essentially martensitic structure, but pure martensite is rare. It is generally accompanied either by small amounts of untransformed (retained) austenite, or by some proportion of primary troostite, which is itself quite hard.

Primary Troostite. What has in the past been called primary troostite, nodular troostite, or troostite of quenching, is now known to be actually a very fine pearlite in which is caught and emulsified not only the cementite and the ferrite of the eutectoid, but also the excess ferrite of the hypoeutectoid steel. It forms when cooling is so fast that the Ar_1 temperature is materially depressed, but not so fast that cementite is not deposited as such and in some approach to a lamellar form. The deposition of cementite at Ar_1 is the characteristic of the Ar_1 change. So, troostite of quenching is the last stage obtained in the size and arrangement of carbide; as the cooling rate is increased, it is the ultimate of fineness and degree of dispersion for directly born carbide.

Demonstration that nodular troostite is merely a pearlitic structure on a finer scale is given by Fig. 4 from the work of Lucas.[1] Nodular troostite in slightly hypoeutectoid steel (similar to the black patches in Fig. 6-b), at 200 X, Fig. 4*a,* shows the dark-etching structure. At higher magnifications, by using a special lens that brings out the definition more clearly than does the ordinary equipment, the true structure is revealed. At 1000 X, Fig. 4*b*, one begins to suspect that the structure is like that of pearlite, and at 4000 X, Fig. 4c, this is definitely proved.

Vilella, Guellich, and Bain[2] are, as previously stated, concerned by the fact that the term troostite is also applied to secondary troostite, a fine non-lamellar dispersion of carbide in ferrite born from martensite on reheating. They suggest that nomenclature would be clarified were this double use abandoned and the term, if used at all, applied only to the product of decomposition of martensite.

Terminology. Something of a dilemma appears, because primary troostite is a structure not attainable in ordinary sizes of C steels without a quench, and when it does appear it is a hair trigger operation to secure it all by itself (except by the expedient of "austempering"). It is normally accompanied by some martensite; so we tend to think of primary troostite as a result of quenching speeds of cooling and as more or less a distinct entity, separate from pearlite. The critical point or change attending its formation has been designated as Ar′ (Ar "prime"). Nevertheless, it is now clear that there is no distinction between Ar_1 and Ar′. They both represent the same change and their products differ only in degree, not in kind. However, at Ar_1 excess ferrite or cementite has been rejected and pearlite left, whereas at Ar′ excess ferrite or cementite is emulsified with the pearlite so that pri-

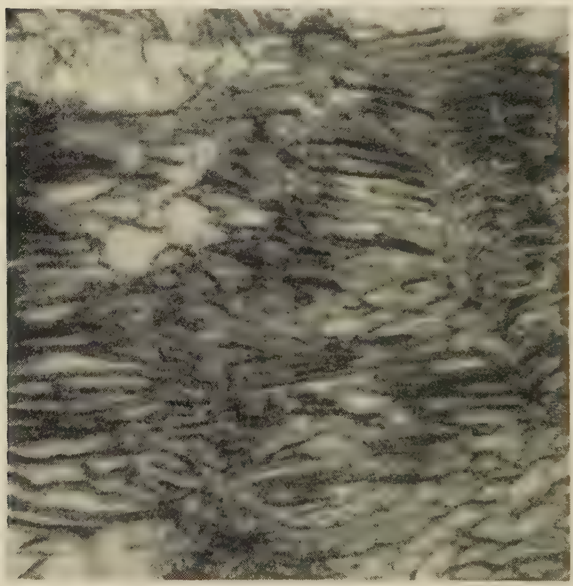

(c) × 4000

Fig. 4 (Continued).—Primary troostite at still higher magnification. (Lucas.[1])

mary troostite patches are not of eutectoid composition as a slow-cooled pearlite patch is. The degree of agglomeration of the cementite and ferrite in primary and secondary troostite is quite similar. Both structures lack the separation of excess ferrite or cementite from patches that are of, or approach, eutectoid composition.

Thus, some other term that might be coined to modify the word pearlite to express the troostite structure would need to carry the distinct implication that the structure referred to need not be (indeed would only be in the case of 0.85% C steel) of, or approach, the eutectoid, pearlite composition. We shall herein continue to use the term troostite, but in using it to designate the structure born directly from austenite, shall call it primary troostite. The nature and source of the structure so designated can be recalled by remembering that *primary* troostite is formed as austenite cools through Ar *"prime"* (Ar′).

The "Hard Constituents." The hardness of a hardened high-C steel in which a little austenite is preserved does not fall very far short of the hardness of the same steel when fully martensitic, probably because the austenite patches are so small. On the whole, however, austenite may be regarded as being considerably softer than martensite, and also much tougher; austenite as obtained in high-Mn and high-Ni steels is but moderately hard.

Martensite, generally regarded as a solid solution of C or carbide in alpha Fe, is the chief and characteristic constituent of hardened steels when cooled rapidly from a temperature above the critical range. Of the transition constituents—austenite to sorbite—martensite is the hardest and also the most brittle, having extremely high tensile strength with little ductility. Microscopically, martensite is usually characterized by a needle-like structure.

There are two definitely distinguishable martensites. The one formed on quenching (at least on drastic quenching) has a body-centered tetragonal crystal structure and is called alpha martensite. On heating only to 210°, or perhaps on just standing, this changes over to beta martensite, the body-centered cubic structure. The two varieties can be differentiated by X-ray crystal structure determinations.

For the purpose of this book we can ordinarily dispense with the distinction between the two martensites, since the alpha form is a transient structure with which we are normally not concerned. We shall need to refer to it only under magnet steels. (Chap. 24, Vol. II.)

Troostite, whether it is the primary troostite formed directly from austenite, or the secondary troostite resulting from a tempering or breaking down of martensite, is intermediate in hardness between martensite and sorbite.

The relative hardnesses of the microconstituents of an 0.85% C steel at room temperature are about as follows:

	Brinell Hardness
Ferrite	90
Austenite—Undeterminable, perhaps about	150
Normalized coarse pearlite	175
Normalized finest pearlite	300
Primary troostite	450
Martensite	675–725
Cementite	650 (approx.)

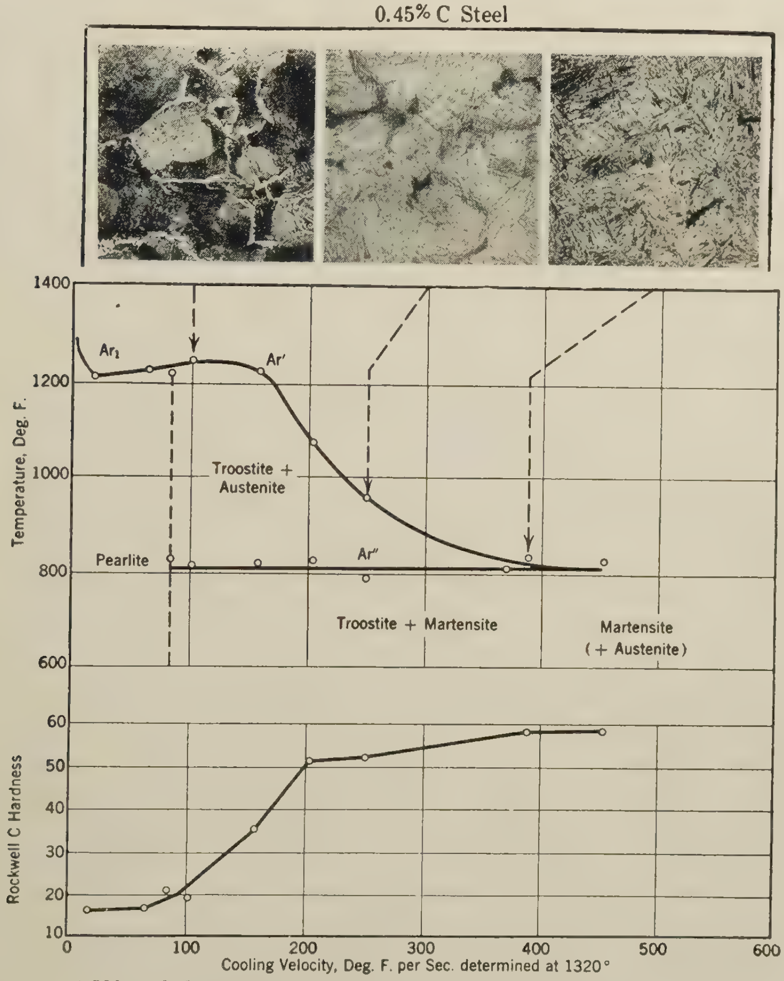
Above about 0.80% C, the hardness of the martensitic steel does not sensibly increase with C content.

Formation of Primary Troostite and Martensite. These transitions, as obtained by progressive increase in the rapidity of cooling in the range about Ar_1, may be illustrated by the quenching diagrams of French and Klopsch[3] shown in Figs. 5 and 6. As the cooling rate increases, Ar_1 is at first gradually and slowly lowered. With further increase in rate of cooling, *two* heat effects appear at widely different temperatures. The upper one, called Ar' (Ar prime) and coincident with the formation of primary troostite, occurs at relatively high temperatures, whereas the lower point, Ar'' (Ar second) at which martensite is formed, is observed at about 700–800°, depending upon the C, Mn, and other alloy contents. With still further increase in cooling rate, the lower point increases in intensity but remains at approximately the same temperature; the upper point decreases and at the same time is lowered somewhat. Concurrent with these heat changes the amount of martensite increases with increase in Ar'', and the troostite decreases. With sufficiently rapid cooling *only* the low point occurs and a martensitic steel is obtained. This is the *critical cooling rate* for complete hardening, with entire absence of primary troostite. Still further increase in the rate of cooling will tend to suppress the occurrence of Ar'' and preserve some austenite unchanged along with martensite; whereas under the most drastic conditions in highly alloyed steels, the complete retention of austenite may be obtained.

The Ar_1 change is, as has been pointed out previously, the same as the Ar' change, but Ar'' is a distinctly different one and occurs at a much lower temperature.

Cooling Rates vs. Structures. Cooling curve 2 in Fig. 7 shows both points, as changes in direction of the curve and a steel cooled at this rate would have a mixed structure, say, like that of *B* or *C* in Fig. 6, whereas curve 1 would have the completely martensitic structure of *E* in Fig. 6. Curve 3 would, of course, have a fine pearlitic structure.

When both arrests are present, as in curve 2, and a structure with both primary troostite and martensite results, it is said that we have a "split transformation." The term is well entrenched but not accurate, for it implies that one is split into two and that Ar″ is of the same nature as Ar′, which is not correct. "Doubled transformation" would be better. We can, however, retain the term "split" if we consider

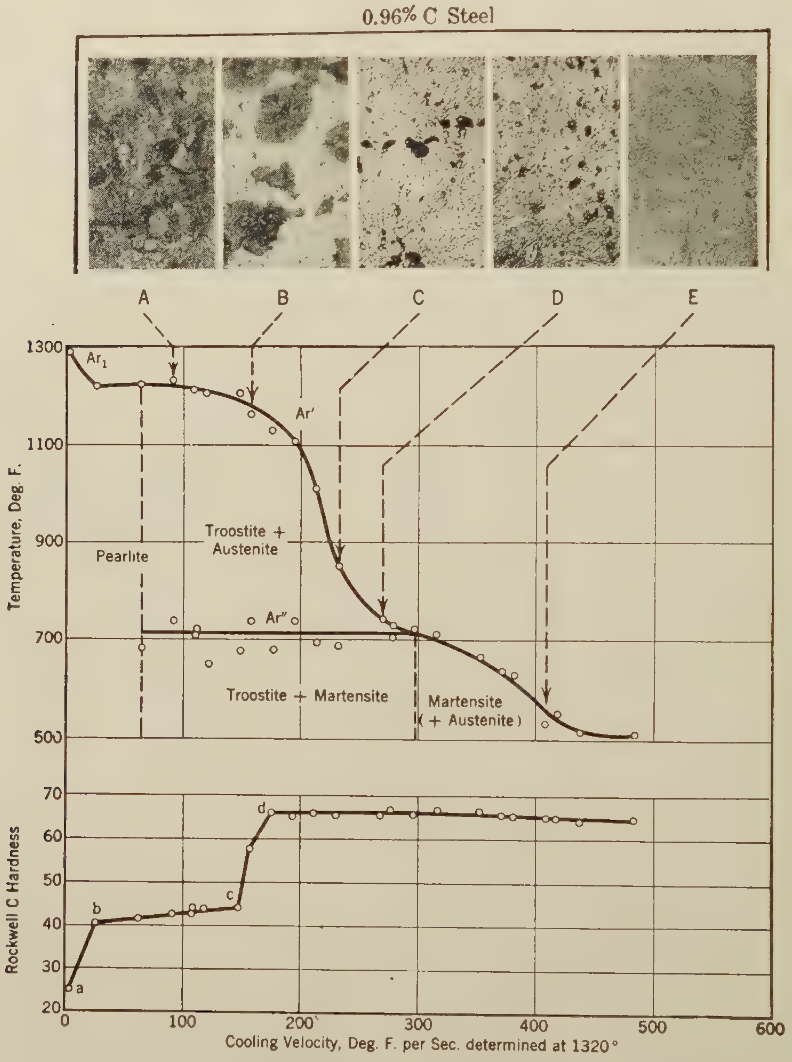


× 500, etched with 2% nitric acid in alcohol. (French and Klopsch.3)

FIG. 5.—"Quenching diagram" for 0.45% C steel, quenched from 1600°.

NOTE: The retention of austenite is shown in Figs. 5 and 6 only for rates higher than the critical. The amount retained increases with the velocity of cooling, and it should therefore be recognized that some may be present in steels cooled somewhat more slowly.

that it means the original austenite is split or divided between two different actions, one occurring at a high, the other at a low temperature. It is the austenite that is unused at Ar′ which transforms at Ar″. Only one transformation occurs at any one microscopic location in the steel.



× 500, etched with 2% nitric acid in alcohol. (French and Klopsch.[3])

Fig. 6.—"Quenching diagram" for 0.96% C steel, quenched from 1600°. See note under Fig. 5.

This discontinuous nature of the propensity toward transformation of austenite has interesting implications. If we cool at the rate of curve 1 in Fig. 7 and once get below the Ar′ temperature by this sufficiently rapid cooling, we have exercised birth control on the austenite. It now has little urge to give birth to pearlite or primary troostite. As DeLong and Palmer[4] put it, the austenite has "given its promise to harden," and on further cooling it becomes eager to give birth to martensite. If the temperature is now dropped to Ar″, martensite will form. If we take it down very slowly and if the steel is one in which Ar″ is not too low, this formation temperature of martensite may be high enough so that, on holding the steel there, the martensite can temper and the finally cooled product will be one or more of the decomposition products of martensite. Or, by continuing the quench, we can retain the martensite.

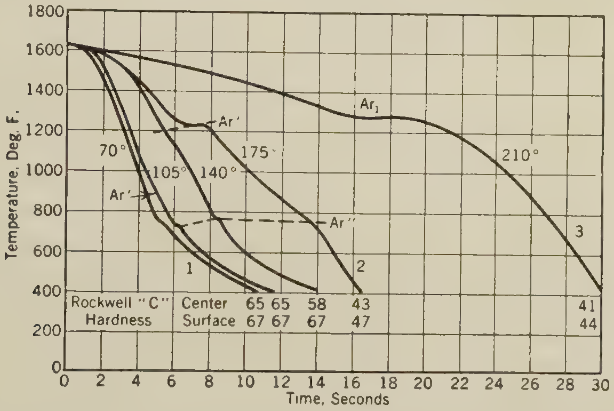


FIG. 7.—Time-temperature cooling curves taken at the center of ½ in. diameter cylinders of 0.95% C steel when quenched from 1605° into still water, with the water at the temperatures shown. (French and Klopsch.[3])

Esser, Eilender, and Spenlé[5] have determined the cooling rates for C steels by means different from those of French and Klopsch. Although there are minor differences in the results owing to experimental difficulties, the general picture is the same. Esser, Eilender, and Spenlé's average results are shown in Fig. 8. The exact temperatures are not so closely determinable as these curves indicate, but instead of plotting the band within which their results scatter, only the averages are here shown.

The position of Ar″ in *B* and the gap in Ar′ in *C* and *D* show the greater ease of hardening steels of eutectoid or hypereutectoid composition compared to those of lower C contents.

Digges [6] has determined the critical cooling rates of straight Fe-C alloys (not steels), all the alloys being of grain size No. 4 (see Chap. 9) after heating to 1700°, and hence are comparable. His data agree more closely with those of Esser, Eilender, and Spenlé than with those of French. Digges thinks that French's specimens were not heated high enough to dissolve all the C in the higher C steels. The curves of the three investigations are combined in Fig. 9.

The question of critical cooling rates for steel of varying C content has also been re-examined by Wever and Rose.[7]

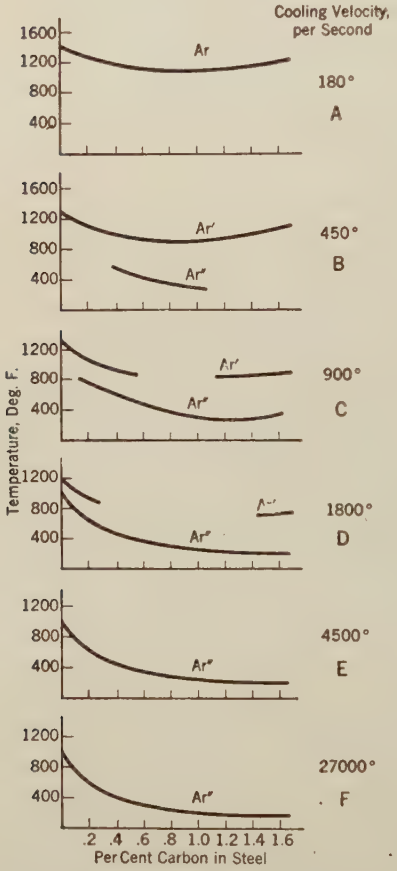


FIG. 8.—Positions of Ar_1, Ar′, and Ar″ at different cooling rates. After Esser, Eilender and Spenlé.[5])

The S-Curve. By cooling below Ar′ and holding at any desired temperature, the propensity toward formation of martensite can be studied. Davenport and Bain [8] have quenched thin samples into molten lead or other low temperature baths, held them there the desired length of time, then quenched in water to fix the structure reached in that time. The results give the S-curve of Fig. 10. The logarithmic time scale of Figs. 10 and 11 should be noted.

The nearness to the zero time scale shows the eagerness with which austenite, once preserved to a temperature below Ar_1, transforms. In the eutectoid C steel of Fig. 10 we see that at 950–1100° austenite wants very badly to transform. If we evade the Ar′ transformation and cool so rapidly that the austenite slides down the left-hand side of the chart without getting ino the "S" zone till it reaches 400° or below, we reach Ar″ and martensite results. If the cooling curve cuts the S-curve at two points, we get both Ar′ and Ar″, the mixed primary troostite-martensite structure resulting. The information given by the S-curve can be collected into plots of the type of Fig. 11[9] which again give us the Ar′ and Ar″ transformations.

Structures Are Characteristic of Their Formation Temperatures. The structures, coarse lamellar pearlite, emulsified pearlite, primary troostite, and martensite, produced on cooling, are pretty much characteristic of the temperatures at which they are born. So if we increase the sluggishness of austenite by adding alloying elements to the steel, we can use a medium cooling rate that in a C steel would give a pearlitic structure, but this time get a martensitic structure, or we can use a much slower cooling rate and get a pearlitic structure in the alloy steel. By holding the C the same and juggling the alloy content and the cooling rate so that the transformation or transformations occur at the same temperature as in the C steel, we get similar structures and approximately the same amount of carbide thrown out. The mechanical properties likewise tend to be quite similar.

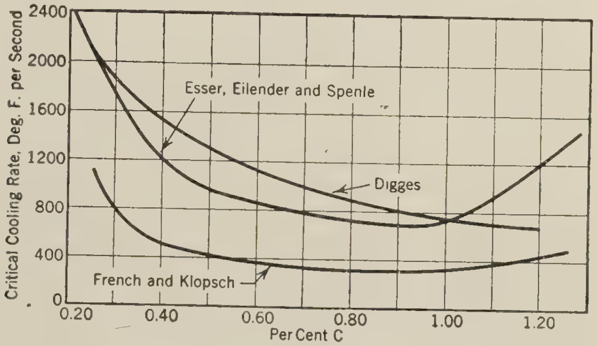


FIG. 9.—Variation of critical cooling rate with C content, as determined by different investigators.

Similarity of Carbon and Alloy Steels. Hence we should not expect the alloy steels to show vastly different properties from C steels; the purpose of the alloy content is primarily to make the steels more easily handled by the heat treater. One advantage of the alloys is that they can be added to low-C steels and confer upon these steels the ability to be cooled at such rates that, on normalizing, Ar′ will occur at a sufficiently depressed temperature to give emulsified pearlite, or, on quenching, to avoid Ar′ and harden completely at Ar″.

The S-curve of Fig. 10 is shifted by C and by alloy content. It can, by alloying, be displaced downwards as well as sideways so that the martensite transformation occurs at a lower temperature range than in plain C steels.

Cold Treatment. Luerrsen and Greene[10] point out that some highly alloyed steels form a borderline class, having sufficient alloy content to depress the transformation below room temperature, but not

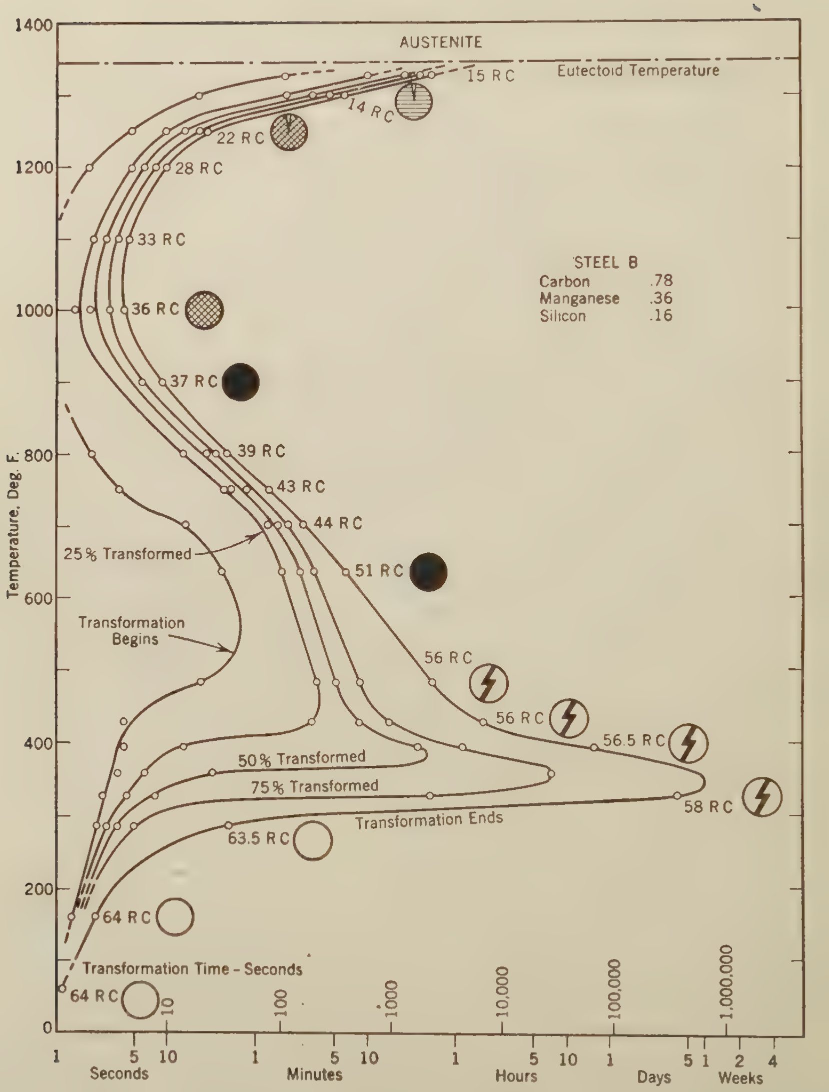


Fig. 10.—Eutectoid steel held at constant sub-critical temperatures. (Davenport and Bain.[8]) Hatched circles represent pearlite; solid circles, primary troostite; zig-zag, primary troostite plus martensite; open, martensite.

Note: Recent data somewhat modify the lower part of Figs. 10 and 11, representing martensite formation. Such data, with S curves for 47 steels, appear in Atlas of Isothermal Transformations, U. S. Steel Corp., 1943.

enough to prevent transformation at some lower temperature. If such a steel is cooled in "dry ice" or liquid air, the steel should be changed from austenite to martensite.

A steel of 0.08 C, 17.15 Cr, 5.06% Ni, quenched from 2000°, cold treated at —112°, had its Brinell raised by the cold treatment from 275 to 445, i.e., it was partially martensitized. The steel was not softened on reheating at 750° but was slightly softened at 930°.

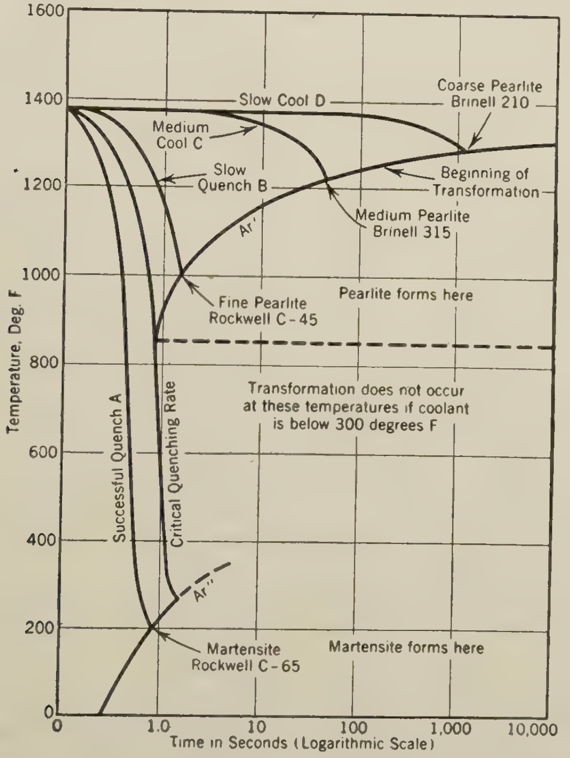


FIG. 11.—Effect of rate of cooling on structure and hardness. U. S. Steel Corp.[9])

With the 750° draw, a tensile of over 200,000, a yield (.01%) of 100,000, 20% elong., 64% R. A., and 80 Izod were secured, which is a marked improvement in properties over those obtainable without the cold treatment.

Of course, only a limited number of steels are susceptible to such treatment, but it is worth while to recall that all possibility for hardening does not cease when steels of this type are cooled to room temperature.

Limitations of Cooling Rates. Even at the highest attainable quenching rates, plain C steel will not harden deeply; i.e., heavy sections will change to martensite only on the outside, the deeper layers will be either troostitic or pearlitic, depending upon how fast they cool. We cannot extract heat from the interior of a steel as rapidly as we can from the surface; the steel will transmit its heat only just so fast. If we could abstract heat at a rapid rate by some analogue of the way we can put heat in by passing electric current through, and generating heat inside, the metal, heat treating would be revolutionized. But, since we cannot create cold in the way we create heat, we have to cool things by letting the heat flow out through the surface to some colder medium.

However, we can influence the cooling rate and hence the structural changes that occur, up to the greatest practically possible rate of extraction of heat. Instead of evaluating the critical cooling rate in degrees per second, it is usually phrased in terms of the "mass effect," the thickness of section that will completely harden under given cooling or quenching conditions. Or, for a given section, a steel is referred to as water, oil, or air hardening, depending upon the cooling medium that will abstract heat at the necessary rate.

Hence the cooling properties of the different quenching media are of interest and will be discussed later in this chapter.

Factors Affecting the Critical Cooling Rate. From the curves of Figs. 8 and 9 one would assume that the critical cooling rate of a C steel was dependent only upon its C content. It is dependent mainly upon the C, but not wholly so. Influences that affect the homogeneity of austenite in turn affect its decomposition. It has been pointed out that whereas all ferrite changes to austenite on heating through Ac_3, the equalization of the C content goes on slowly so that considerable time or a higher temperature is needed to produce homogeneous austenite. Until equalization is complete, the edges of the austenite grains are relatively low in C. As Fig. 8 shows, low-C steels have high critical cooling rates; so at a given fairly slow rate of cooling that would, in homogeneous austenite of the average C content, produce a fully hardened structure, the low-C edge will not cool through Ar_1 and Ar' without transformation to harden later but will instead go over to pearlite at Ar_1 or primary troostite at Ar'. Once this action is started it proceeds through the whole grain. Unhomogenized austenite, therefore, is active, and the trigger action of the edge sets off the whole. Fully homogenized austenite is far more sluggish and can be more readily cooled through Ar' without transforming. This was the difference between Fig. 2*d* and 2*e*.

Effect of Maximum Temperature. In some alloy steels this behavior is so marked that a quite definite temperature of austenite can be determined, below which the austenite is incapable of hardening at a given cooling rate, but at or above which cooling at that rate results in a clear doubling of the transformation to $Ar' + Ar''$ and the appearance of hardening corresponding to Ar''. Swinden * noted this in W and in Mo steels, and the temperature in question was denoted "T max.," i.e., the maximum temperature for avoidance of hardening. Howe, Foley, and Winlock [11] showed the same thing to be true for C steels. Ellis [12] discussed it in detail for S.A.E. 3130 Ni-Cr steel.

The effect of quenching temperature on the cooling velocity through

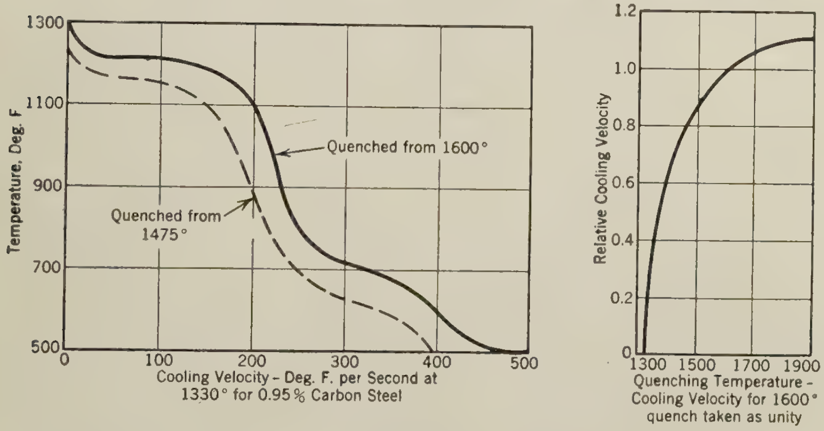


FIG. 12.—Effect of increase in quenching temperature on cooling velocity. (French.[13])

Ar_1 at the center of spheres of high-C steel is shown by French in Fig. 12.

The size of the specimen, the ratio of volume to surface, and the composition all affect the actual center cooling rate, and the original article by French [13] discusses how these factors may be calculated. Without entering into these calculations here, Fig. 12 indicates that *increase in quenching temperature* will affect the critical cooling rate and increase the depth of hardening. Commercially, if the proper temperature for hardening is found by fracture tests on ½-in. diam. specimens, that temperature will be exceeded by at least 50° when maximum depth of hardening is sought on 2-in. diam. areas.

Grain Growth. Connected with this behavior is the observation that grain growth in austenite, which we have heretofore referred to as tending to start immediately above Ac_3 on heating and to increase in intensity with increase in temperature, actually does not go on rapidly until some elevated temperature has been reached. The coarsening temperature depends not so much upon the chemical composition of the steel as to main constituents, as upon the presence or absence of a tiny amount of some so-called "deoxidizing" element, perhaps as oxide or other compounds, perhaps as the element itself. The urge to coarsen is pretty much inherent in the steel, i.e., is fixed as it freezes from the melt. Heat treatment and hot or cold working may modify this urge somewhat, but on the whole a steel of a given coarsening tendency strongly retains that tendency.

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* *Jour. Iron & Steel Inst.*, v. 73, 1907, p. 291; v. 80, 1909, p. 223; *Carn. Sch. Mem.*, v. 3, 1911, p. 66; v. 5, 1913, p. 100.

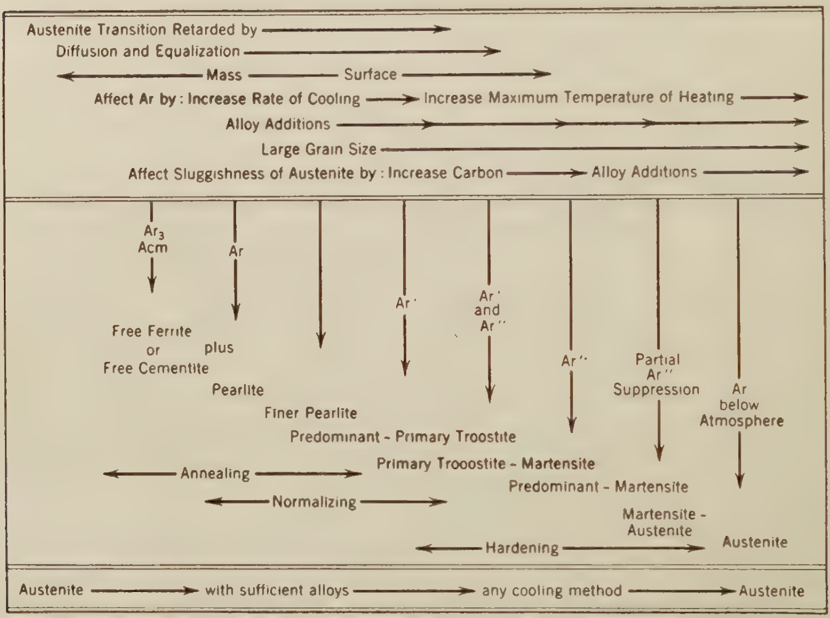

FIG. 13.—Factors affecting the transformation of austenite.

A coarse-grained steel has a marked urge to harden in the course of heat treatment. It is, for its composition, a deep-hardening steel. On the other hand, if the grain size is fine, the urge to harden is less; it is a shallow-hardening steel. The subject of grain size will be more completely discussed in Chap. 9.

Thus we may note that, besides the extraneous feature of rate of cooling, the degree of hardening is affected by the temperature of heating and the grain size.

Hardening Chart. In the present conception of the austenite transition, expressed in the chart of Fig. 13, the mechanics of hardening are explained through the factors governing the retention of austenite or its early decomposition products.

The transition of austenite through Ar_1, Ar', and Ar'' is obstructed or retarded, with cumulative effect, as follows:

- *A*. Physical phase
  - 1. Diminishing mass
  - 2. Increasing surface
- *B*. Heating phase: diffusion and equalization
- *C*. Cooling phase, affecting Ar ranges
  - 1. Increasing rate of cooling
  - 2. Increasing maximum temperature of heating
  - 3. Alloy additions
- *D*. Chemical phase, affecting sluggishness of austenite
  - 1. Increasing carbon
  - 2. Alloy additions
- *E*. Structural phase: grain size

These factors, as has been explained, determine the structural constituents resulting, and these, in turn, define the degree of hardening, or "hardness." Thus, the chart indicates that with retarded cooling, as in annealing, the usual ferrite (or cementite) precipitation occurs between Ar_3 and Ar_1, followed by normal pearlite at Ar_1. A slight retarding of the austenite transition will produce finer pearlite. Further retarding will produce the doubled changes Ar' and Ar'', with Ar'' increasing and Ar' decreasing, followed, in turn—and especially with alloy additions—by the suppression of Ar'' or carrying it below atmospheric temperatures.

Critical Cooling Rate. That cooling velocity which first results in a martensitic steel free from troostite is called the "critical cooling rate." In order to attain this critical cooling rate, that quenching medium must be selected which, under definite conditions of temperature and volume velocity—both referring to the quenching medium—will produce a predominant martensite structure.

Quenching Factors. From the practical standpoint, a discussion of the quenching factors involved must include the following:

I. The quenching media, such as

- *A*. Oils
- *B*. Water
- *C*. Brine
- *D*. Special solutions, such as sulphuric acid, sodium hydroxide, etc.

II. Temperature of the quenching medium.
III. Volume velocity of the quenching medium.

Quenching-media Diagram. The effect of various quenching solutions upon some of the properties of C steels, with comparisons for a fixed size and shape (½ in. round sections), has been investigated by French and Klopsch[3] and liberal excerpts, slightly modified in wording, are here taken from their report.

In Fig. 14 are shown the thermal transformations and structures produced in 0.45, 0.75, 0.95 and 1.25% C steels with varying rates of cooling.

"No single numerical value selected from time-temperature cooling curves will give a quantitative measure of the 'hardening ability' of various quenching media; such comparisons require consideration of the entire cooling curve. However, for all the aqueous solutions and oils tested, the cooling velocity at 1320° permits a general grading consistent with results produced in quenched C steels in ½-in. sections. With this in mind the various liquids studied increase in 'hardening ability' in the following order:

"Water at 210°
"Water at 175° and the various oils at 68°
"Water at 140°
"95 per cent. sulphuric acid at 70°
"Water at 105°
"Water at 70°
"10 per cent. sulphuric acid } at 70°
"10 per cent. sodium chloride } at 70°
"5 per cent. sodium hydroxide } at 70°

"None of these showed extreme sensitivity productive of erratic results when kept in fairly vigorous motion about the steel. However, if the section is such that a cooling velocity close to the critical is produced in the zone which is to be hardened, great difficulty will be encountered in securing consistently uniform results with ordinary methods of heat treatment. This is not extreme sensitivity of the quenching solution but of the steel under treatment in relation to definite cooling rates. Thus where a fully hardened steel is desired at a given zone in the cross section a coolant should be chosen which will give in that zone a cooling velocity somewhat in excess of the critical rate. If the chosen liquid will likewise cool extremely rapidly at the lower temperatures the steel can be removed from the quenching bath before it is entirely cold and so minimize any increase in the tendency to crack. However, it is quite evident that media having 'hardening ability' greatly in excess of that required for the work in view, should not be used.

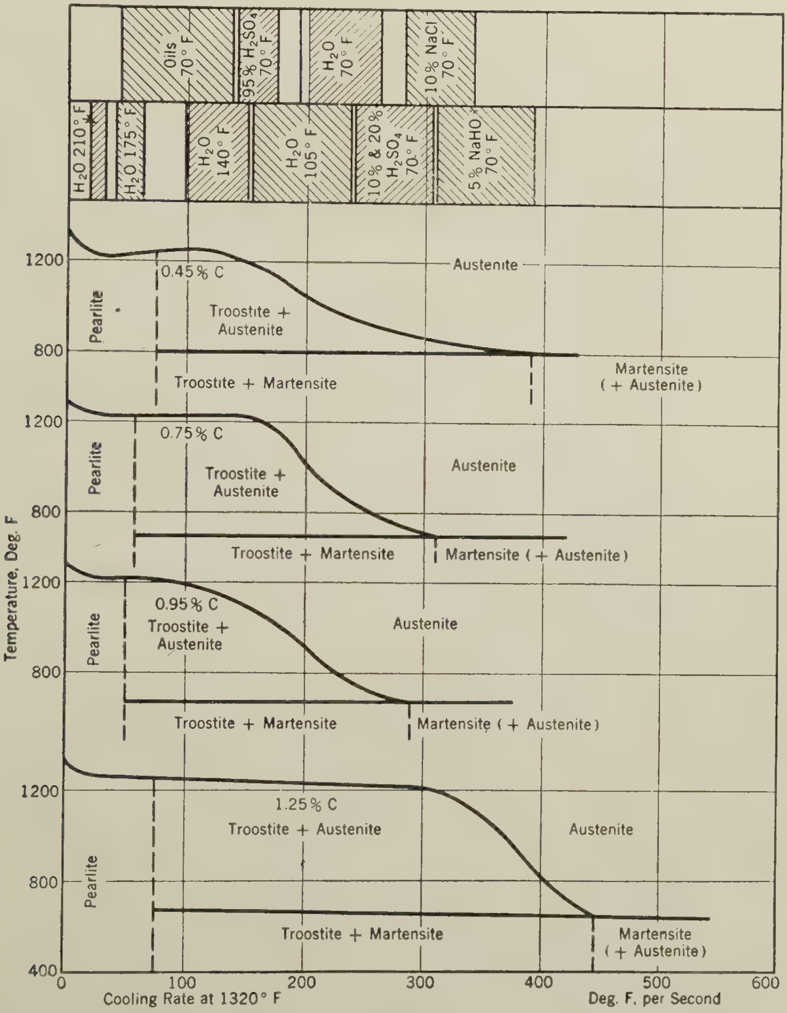

FIG. 14.—Relation between some quenching media for heat treatment and quenching diagrams for carbon steels. (French and Klopsch.[3])

"In addition to a higher cooling velocity in the range which largely determines the degree of hardening in C steels, dilute sodium hydroxide solutions 'bite' more quickly than water, i.e., they begin to extract heat at a rapid rate from the steel almost immediately after immersion instead of showing the definite initial period of comparative 'inaction' of water. * The same characteristic, but to a somewhat lesser degree, is shown by the dilute sulphuric acid solutions and brines and permits the use of lower hardening heats for high-C steels, with advantages attending such procedure. However, these media maintain relatively high cooling rates at low temperatures where the steel is least able to withstand stresses induced by thermal gradients.

Water Quenching. "It is at once evident from Fig. 14 that still water at a temperature just under the boiling point takes away the heat so slowly that in C steels, pearlite results. Still water at 175° is a somewhat more rapid medium and produces pearlite or combinations of pearlite, troostite, and small proportions of martensite depending upon the quenching temperature, chemical composition, and initial condition of the steel under treatment. Still water at 140° and 105° or water flowing past the steel at 1 foot per second at 70° all give a troosto-martensitic product. However, the amount of martensite increases as the water temperature decreases, but in each case it likewise is dependent upon quenching temperature, steel composition, etc.

"It should be kept in mind that the described results were obtained at the center of ½-in. cylinders and that the cooling velocity and, hence, hardening, is much greater at the surface. This is illustrated in Fig. 7, which shows the center and surface hardnesses for the four specimens quenched, as indicated, into water at various temperatures. In no case is a fully hardened center obtained (martensitic steel free from primary troostite), a fact which is also clearly shown in Fig. 14 as the most rapid cooling in the water series is well below the critical cooling rate of all four steels. However, a fully hardened surface is obtained in water at 70° or 105° and closely approximated by still water at 140°.

"The inability to surface-harden sizes as small as ½-in. by the use of water at temperatures above about 140° is not its only undesirable characteristic as a coolant. As is well known and shown for 0.95% C steel in Fig. 7, the most rapid cooling is obtained between 750° and 390° instead of, as in water between 70° and 105°, at higher temperatures where the steel is better able to withstand stresses set up by thermal gradients. Likewise a given change in water temperature between about 105° and the boiling-point has a much greater effect upon cooling velocities than in water between 70° and 105°. The changes in the latter case are very small and show no marked sensitivity to temperature change. However, it should be recognized that with certain combinations of steel composition, size and shape of material and quenching temperature a cooling rate close to the critical may be obtained, in which case small variations in the temperature of the bath or other factors affecting to a relatively small degree its cooling characteristics may produce appreciable differences in the quenched steel."

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* High-speed photography of specimens quenched in water, brine, caustic solution, and transparent oil enabled Zavarine[14] to explain this nicely. In brine or caustic, the heating of the liquid at the surface of the hot piece concentrates it to saturation and causes solid crystals of NaCl or NaOH to form on the steel surface. As this film builds up to a certain thickness it disrupts, explosively, throwing off a cloud of crystals. This discharge of crystals carries away the steam bubbles that cling to the steel surface and tend to blanket it.

Scott[15] states that the rate of heat dissipation by the liquid itself is scarcely different in water and in brine. Hence it is clear that the sharper quenching in brine is really due to effective circulation of the water by the mechanical action of the salt. High-pressure water quenching that is effective in carrying away the vapor jacket will accomplish the same thing as brine.

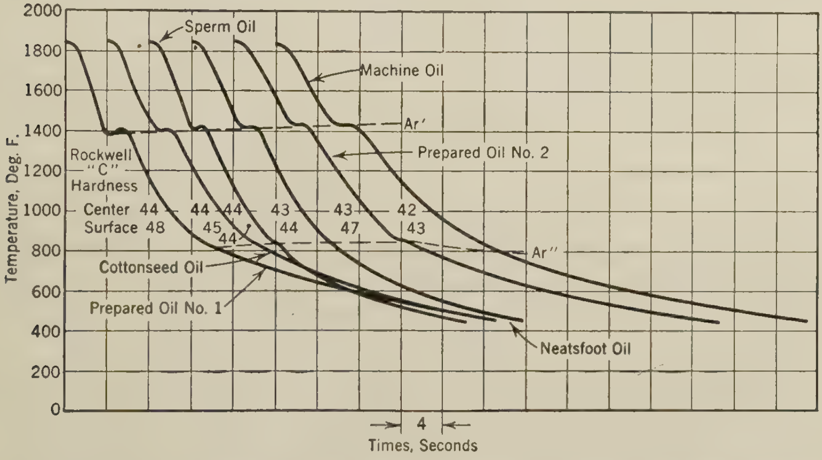
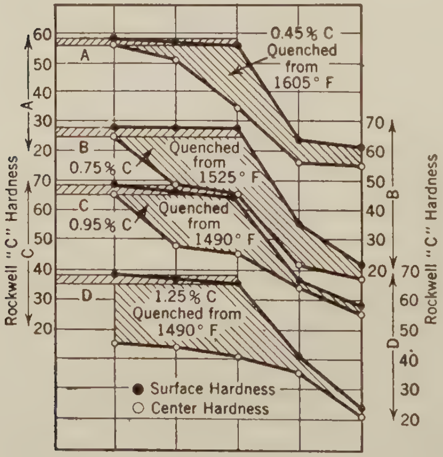


FIG. 15.—Time-temperature cooling curves at center of ½-in. diameter cylinders of 0.95% C steel, quenched from 1605° into various oils motionless at 70°. (French and Klopsch.[3])

Oil Quenching. "Time-temperature cooling curves for the six oils referred to in the table, when used to quench 0.95% C steel from 1605° are reproduced in Fig. 15, and there is likewise given in Fig. 16, for comparison with the quenching diagrams of C steels, the approximate range in cooling rates to be expected with varying quenching temperatures between 1490° and 1605°. The oils all give an extremely slow rate of temperature change below about 570° to 660° in contrast with the rapid cooling in hot water, already described. If, as is the case with the 0.95% C steel the formation of martensite takes place at about 660° to 750° it is probably largely tempered during the subsequent slow cooling. The net result in ½-in. sections of 0.45 to 1.25% C steels is the production of primary troostite with varying but small proportions of martensite, and either free ferrite or cementite, depending upon whether the steel is respectively hypo- or hyper-eutectoid. Thus intermediate hardness values are obtained.

Oil Used	Specific Gravity at 59°	Flash-point, Degrees	Fire-point, Degrees
Prepared No. 1	0.862	360	405
Cottonseed	0.915	310	685
Sperm	0.882	500	590
Neat's-foot	0.919	480	620
Prepared No. 2	0.874	375	415
Machine	0.909	405	405

"The use of oils also results in greater uniformity in microstructure and hardness throughout the cross-section than is the case when using water at temperatures which give similar rates of cooling at 1320°. This is clearly shown by comparison of the Rockwell hardness curves in Fig. 15 with values given in Fig. 16, and, as might be expected from such conditions, a fully hardened surface was not obtained in any case.



Surface and center hardness obtained in ½-in. sections by quenching different steels into water at various temperatures. The upper cross-hatched band for each steel represents the hardness obtained with the critical cooling rate. (The scale used in plotting changes for each steel.)

Fig. 16.—(French and Klopsch[3].)

"The above discussion refers to oils kept motionless at 70°, but variation in temperature at least up to 175° has a very minor effect on the time-temperature cooling curves, so that the described effects apply within a somewhat wider field.

"If, instead of a motionless quenching bath, the oil flows past the specimen at 1.3 ft. per second, a slight acceleration in cooling is observed. This change is relatively unimportant in relation to the foregoing tests, but may cause a considerable change in hardness and microstructure with certain combinations of quenching temperature, size and shape of material and steel composition, which bring the cooling rates within 'sensitive zones' such as those indicated in Fig. 15.

"The six oils tested may be rated qualitatively in order of decrease in 'hardening power' (cooling velocity at 1320°) as follows:

"1 Prepared quenching oil No 1.
"2. Cottonseed oil.
"3. Sperm oil. } Have almost identical cooling rates at
"4. Neat's-foot oil. } 1320°
"5. Prepared quenching oil No 2.
"6. Machine oil.

"While the quenching properties of each of these oils are generally characteristic of the group as a whole and no very wide variations are observed except a rather rapid cooling at low temperatures in sperm and neat's-foot oils, other factors such as permanence in continued use and a high fire or flash point are of considerable practical importance and differences in these latter respects would materially affect the choice of any one oil in the group."

Brine Quenching. "Various concentrations of sodium chloride, between 10% and saturation, in water at 70° (brines) may, as a group, be considered intermediate between 10% sulphuric acid and 5% sodium hydroxide solutions. However, they overlap the range in cooling rates (Fig. 14) for the somewhat slower dilute sulphuric acid and the more rapid sodium hydroxide and give cooling velocities well above water. With favorable conditions with respect to quenching temperature and motion of the bath, it is possible to fully harden to the center of ½-in. round sections, steels from about 0.60 to 1.05% C."

Requirements of Quenching Oils. Quenching oils are primarily evaluated on their ability to dissipate heat from the surface of the steel, but other factors come in also. The flash point must be high enough so that there is no fire hazard. The viscosity should be as low as is consistent with the other requirements, to minimize drag-out. The cooling rate should not be too markedly affected by slight changes in temperature of the oil. The oil should be stable, not cracking and depositing C or gum on the work and not changing with use so as to lose its initial quenching rate.

Proprietary oils are prepared to meet some combination of these requirements, the ways they are blended being held as trade secrets. Large users generally prepare their own oils or buy from the oil companies on specification. Such specifications are individual and have not yet been standardized by national technical associations. In the large-scale heat-treating operations of the automobile companies, mineral oils selected on the basis of experience appear to be most used. A straight run mineral oil of 100 sec. Saybolt viscosity at 100° is rather common.

Addition of about 5% of a fatty oil like lard oil to a mineral oil of suitable viscosity (both at room temperature and at the temperature the quenching bath may acquire) makes a satisfactory quenching oil. The fatty oil must serve much the same purpose as do such additions in lubricating oils, increasing the ease of "wetting" the metal surface. Although it is often feared that such an oil may deteriorate rapidly, some such oils show slightly better cooling power after months of use than at the start, the explanation probably being that the more volatile fractions that tend to form a vapor blanket about the work were driven off in use.

It would be desirable to have available a quenching medium intermediate between oil and cold water. Glycerine solutions, water-glass solutions, oil-water emulsions, and hot water are sometimes used to bridge this gap; but there are practical difficulties in their use, and the problem is still not satisfactorily solved.

Anything that tends to form a film on the work which prevents good contact, or causes irregular contact between the work and the coolant, is harmful. A little soap in water used for quenching leads to irregular contact. Quenching conditions such that a bubble of steam or of oil vapor adheres to the work will cause a soft spot beneath the bubble. In quenching by hand the work is swashed about, and in mechanical quenching the coolant may be pumped or sprayed so as to keep it in rapid motion and prevent trapping of bubbles.

Effect of Nature of the Surface. A peculiarly difficult problem is the hardening of work so scaled in the heating that on immersion the scale cracks off in spots, as the bare spots harden whereas those protected by scale may not. Brine or sodium hydroxide quenches tend to cause the scale to snap off immediately upon immersion.

Sometimes the mechanical condition of the surface of the steel, whether it is rough ground, polished, electroplated with some metal, or what not, may so affect the wettability of the steel by the coolant, and the evolution of gas bubbles or the retention of an insulating gas film, as to have a material effect upon the behavior in hardening.

Herty, McBride, and Hollenback[16] examined fine-grained S.A.E. 1045 steel quenched after heating in an oxidizing atmosphere, which produced a heavy scale, and in a "neutral," controlled atmosphere, which produced only a very light scale. Some 3000 Rockwell readings were made on the surface of comparable specimens. With the heavy scale 8½% of the readings showed incomplete hardening—i.e., soft spots, against 1% with the light scale.

Practical Considerations in Quenching. The foregoing discussion has dealt with small masses and has indicated in a technical way the salient features of various quenching media. In commercial work, however, there must not be lost to sight the practical viewpoint which affects the uniformity and therefore the cost of the finished hardened product. This involves a consideration of the following factors:

1. The temperature and volume of the bath;
2. The volume velocity of the bath;
3. Hardening temperature *vs.* cooling rate; and
4. The method of handling material to, through, and from the bath.

Temperature Volume of Cooling Medium. In order to obtain uniformity of cooled product in accordance with the technical factors determined in the laboratory, there must first be established a uniformity of temperature of the quenching medium in relation to the definite mass-surface factors of the material and production required. Due consideration under the distinctive conditions must be given to the heat-absorptive power of the liquid, its capacity for transferring the heat thus absorbed to the cooler part of the bath, the viscosity as affecting the motion of the liquid, and the volatility as influencing the formation of a vapor around the steel, thus preventing a uniform removal of heat. Circulation must be provided to take away the hot liquid and replace it with liquid at the temperature required. A suitable cooling system—mandatory with oil—must be installed to insure a sufficient volume of cold quenching solution to maintain the uniformity of temperature in the bath. And, finally, just as much skill and ingenuity must be used to provide that uniformity in temperature time and rate of cooling as were given to the same factors in heating.

Volume Velocity of Cooling Medium. The preceding paragraph assumes a quenching bath of standard temperature volume, i.e., a flow of the cooling medium only sufficient to maintain a uniform standard temperature. It is under these conditions that the greatest amount of hardening work is done. On the other hand, there are frequent and important instances in which the rapidity of cooling required to produce the metallurgical results is greater than that obtainable under the usual temperature-volume conditions. Under such conditions the velocity of the cooling medium can be increased by greater flow (volume velocity), with or without localized pressure.

Pressure Quenching. Water sprayed under pressure against the particular part of the piece to be hardened is a very common example of localized pressure application. It is the quickest agent for rapid cooling in common use, exceeding in its hardening ability both brine and still water baths. The main point to be noted is that there shall be sufficient volume and pressure to prevent the formation of a blanket of steam between the hot steel and the spray. The most common use of the water spray is for differentially hardening such work as die-block faces, anvils, sledges, rivet sets, etc. Many ingenious devices for localized hardening have been worked out, as in the hardening of extruding dies,[17] which presents interesting problems.

Very rapid cooling of comparatively small pieces, such as rivets, bolts, etc., can be obtained by allowing the hot steel to drop through a pipe or other passage to which the cooling medium is supplied continuously at high volume velocity. Smith [18] has shown that it is possible by such methods in commercial production to obtain tensile strengths as high as 160,000 from 0.09 to 0.15% C steels.

The modern gear-hardening machine is another noteworthy example of volume-velocity quenching, involving not only a high rate of cooling but also the reduction of the distortion factor. The gear is clamped tightly between dies through which the quenching oil is forced under pressure. The process is important on account of high physical test results, marked uniformity, and absence of warping.

The effect of cooling rates, quenching media, and quenching temperatures was nicely demonstrated by Jungbluth [19] who quenched small pieces of a 0.50 C, 5% Ni steel, which could be made to retain practically all its austenite by rapid quenching from excessively high temperatures, with the results shown in Fig. 17.

Hardening Temperature *vs.* Cooling Rate. There is no distinctive hardening temperature suitable for all conditions. Customary practice suggests a temperature slightly in excess of the end point of the upper critical on account of the maximum grain refinement and formation of the solid solution there occurring. But it should be evident from the previous discussion that there are many other factors to be considered, all of which have a bearing on deciding the issue. On the one hand, the shape, size, and physical characteristics of pieces to be hardened, the chemical composition, and the character of the initial structure are fixed and definite; on the other hand, certain results are desired. Between the two there lies a distinctive cooling rate through the critical ranges in relation to the cooling medium which determines the temperature of heating and completes the cycle.

Handling Methods. From the practical as well as from the technical viewpoint, close study should be given to methods of handling the material to, through, and from the cooling medium. Regardless of the temperature indicated in the cooling bath, the pieces will not be cooled alike unless exposed in the same manner, at the same rate, for the same time, at the same temperature, and in the same atmosphere. The economic, as well as technical, importance of the cooling cycle will be developed more fully in Chap. 17.

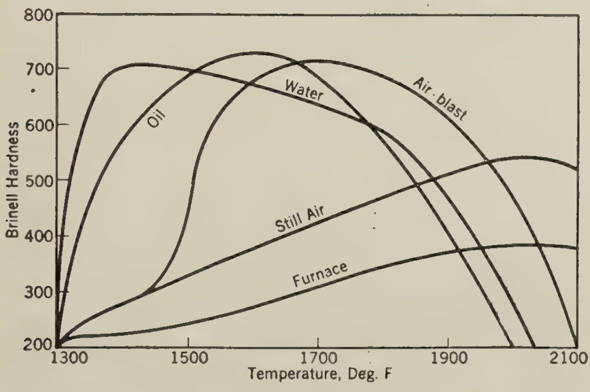


FIG. 17.—Hardness of 0.50 C, 5% Ni steel quenched or cooled in various media from the temperatures shown. (After Jungbluth.[19])

Time Quenching. So-called "time quenching," i.e., holding a large piece or a certain part of a piece in the quenching medium for a specified length of time until the outside of the immersed part has cooled through Ar″ to form martensite, then withdrawing it and allowing the stored heat in the interior or in the unquenched body of the piece to flow into the hardened part and thus temper it in the one operation, is an operation difficult to control. Fassinger[20] examined the process and decided that it was definitely inferior to a full quench and temper. Specific instances may arise where the method may serve.

Dimensional Changes and Internal Stresses Resulting from Quenching. At the proper temperature for full hardening by quenching, steel is completely austenitic. If it could be cooled without the austenite changing to any other constituent it would shrink gradually and uniformly as it cools, by about 0.000013 in. per inch per deg. F.

But pearlitic hypoeutectoid steels shrink only about 0.000008 in. per inch per deg. F. On slow cooling through Ar, steel expands, and conversely in heating up through Ac, it contracts. Austenite is more dense than pearlite or ferrite.

We have seen that the transformation goes on at different rates at different speeds of cooling so that the outside of a piece of steel cooling through Ar tries to pull away from the inside which is slower in cooling. In heating through Ac, the austenitic exterior exerts pressure on the untransformed pearlitic interior and, if this pressure is exerted too rapidly or in unbalanced fashion, in a piece of irregular contour or section, the outside may crack; hence the necessity for care in heating. In a non-symmetrical section or in an unevenly heated or cooled piece, dimensional changes tend to produce warping.

Volume Changes. When austenite transforms to martensite at Ar on quenching, the steel expands, for martensite is less dense than austenite. This does not mean that a piece always grows longer; it may actually shorten a trifle, but when it does it grows enough fatter so that the net result is expansion. The lengthening, fattening, or combination of both, depend on the dimensions of the piece, and the depth of hardening. The changes are hard to predict mathematically and are most readily found by trial. Since large pieces do not harden clear through and since martensite is brittle, this expansion on quenching tends to spall off the outer layer, or to cause cracking to relieve the stress caused by this pull of the outer layer on the underlying metal. Hence it is customary and often essential to start tempering a hardened piece before its gets completely cold, since, given a little time at room temperature, the piece may crack. This may be due to the fact that alpha martensite (tetragonal) which exists at the lower temperature transforms to beta martensite (cubic) at about 210°, and the alpha may be a more brittle form. It may merely be a result of the stress distribution changing only slowly with time on account of the rigidity of the steel at these low temperatures. Untempered martensite, not stress relieved, is likely to crack on grinding. Carburized parts are mildly tempered to minimize grinding cracks.

Effect of Tempering. As martensite is tempered it shrinks gradually until the piece finally gets back to practically the same dimensions it would have had had ferrite and pearlite been formed by slow cooling the austenite instead of quenching and tempering. But, since hardness requirements seldom permit full softening, a quenched and tempered object will ordinarily have different dimensions after heat treatment than before. In making dies, reamers, gears, etc., these changes have to be taken into account. The tendency toward dimensional changes, warping, and cracking may be studied by the use of a specimen shaped as shown in Fig. 18, which is measured, quenched, and tempered, and again measured. The size of the specimen may vary according to the degree of precision desired.

Retained Austenite. Some austenite (traces to a few per cent) may be retained on quenching even in C or slightly alloyed steels, and with very high alloy content the austenite content of the quenched steel may be high. When a little austenite is retained with the martensite it cuts down the expansion on quenching proportionately. On tempering, the retained austenite transforms. In some special cases of high-alloy steels and "cold treatment," e.g., in liquid air, the retained austenite may be changed to martensite. Ordinarily, in a steel quenched so as to be chiefly martensite with some austenite, the austenite remains as such until the temperature is such as will begin to transform the martensite to secondary troostite, i.e., above the temperature at which martensite is stable. The dimensional changes attending this tempering alter the pressure on the austenite and it responds by changing to secondary troostite, as it would change at that temperature (as in an austempering process) if it had not been trapped on cooling.

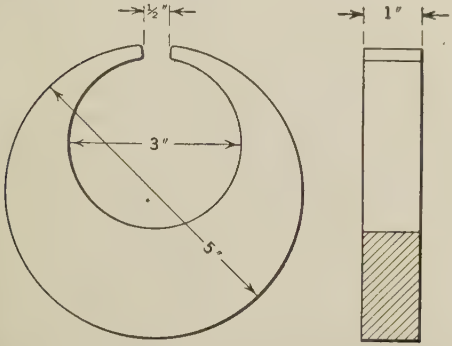


Fig. 18.—Type of specimen used for study of dimensional changes on quenching. The size of the specimen may be varied, and the specimen quenched once, or repeatedly, according to the information desired.

Thus a quenched martensitic steel containing troostite will soften slightly on low temperature tempering, but at a slightly higher temperature will increase in hardness owing to replacement of soft austenite by harder primary troostite. This is one type of "secondary hardening"; another type is ascribable to precipitation hardening.

Non-deforming Steels. It may be possible to balance the composition in such a way that the expansion on quenching is so reduced by the presence of austenite that the expansion is so slight as to be balanced by the scaling in heating for quenching with the result that the quenched and slightly tempered piece is essentially of the same dimensions as before hardening. "Non-deforming" tool and die steels result.

Stresses from Volume Changes. All the steels in transforming from austenite to martensite or from martensite to ferrite and pearlite or cementite and pearlite go through dimensional changes of considerable magnitude, and at temperatures where the steel is quite rigid. The transformations do not occur simultaneously throughout the steel so that local "pushes and pulls" are exerted. The steel is too strong to yield readily, so internal stresses are set up whose magnitude and distribution vary with the object treated and with many of the attendant circumstances. Calculation of these stresses in a particular case may lead into complicated mathematics. Eaton [21] has given a qualitative discussion, incidentally pointing out that a decarburized skin complicates the stresses exerted, because it does not harden fully at the cooling rate required by the underlying metal.

Fortunately, in the tempering process the previous dimensional changes are somewhat counteracted and at the higher tempering temperatures the steel is not too rigid to allow redistribution and equalization of the stresses. Fortunately, too, if long tempering time is given even at fairly low temperatures, gradual relief of stress occurs, by internal mechanism perhaps somewhat analogous to creep at high temperatures.

It is because of this ready relief of stress at high temperature that the comment is justified that, other things being equal, the steel that gives the properties desired at the higher tempering temperature is the one to choose, for without doubt internal stress lowers the endurance and the impact properties. The deviation of the endurance limit from the 50% ratio (Fig. 14, Chap. 1) is most probably related chiefly to the internal unrelieved quenching stresses in the very hard, only lightly tempered, steels (and to the propensity toward formation of microscopic cracks in martensite needles during quenching, as has been discussed under "austempering").

Avoidance of Unbalanced Stresses in Quenching. The effect of unbalanced stresses in quenching is clearly shown by Fig. 19.[22] A hollow cylinder was to be quenched by flushing the bore only, but some water was spilled on the outside, with resultant irregular hardening and cracking. When precautions were taken to avoid spilling so that the bore only was hardened, and very symmetrically, there was no cracking because the stresses were balanced.

The stresses produced can be estimated by grinding off layers, measuring the changes in dimensions, and calculation, on the basis of the modulus of elasticity. Greene[22] thus found internal stresses of the order of 125,000 to 250,000 lb./in.^{2} tension or compression.

A little thought as to the effect of internal stresses that are not evenly balanced will emphasize the usual precautions as to avoiding drastic quenches on irregular bodies, and show why a cylindrical rod to be hardened is quenched vertically rather than horizontally.

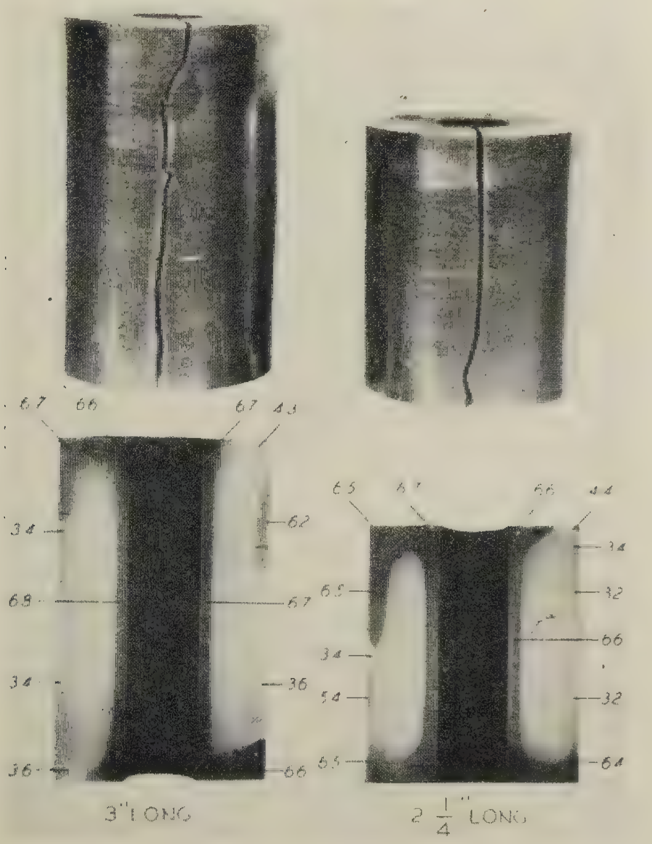


Fig. 19.—Cracking resulting when, in an attempt to flush the bore only, in a C tool steel of 0.91 C, 0.25 Mn, 0.18 Si, 0.013 P, 0.016 S, 0.07 Cr, 0.10% Ni, water was spilled on the outside also. The steel was flushed at 1600°. On cooling to room temperature, the cylinders cracked as shown at the top. The variable penetration of hardness and the Rockwell C hardness numbers obtained are shown, on sections, at the bottom. (Greene.[22])

If the quenching stresses are balanced, as they are in a properly quenched cylindrical object, quenching cracks can be avoided. At first sight the method of surface hardening of crankshafts, piston pins, and the like by induced electric heating followed by spray quenching would seem difficult to accomplish without cracking. In this process the surface is heated so rapidly that the steel is heated *only* on the surface. For example, if a piece 1¼ in. round × 1⅞ in. long is to be hardened, 130 kw. are applied for only 4.3 sec. The electric energy is induced in the piece itself by surrounding it with a single-turn primary coil carrying high-frequency current. The instant the required amount of energy has been absorbed, the power is automatically shut off and high-pressure water spray applied through perforations in the primary coil. (See Fig. 15, Chap. 13.) A low temperature stress relieving draw is given afterwards. In spite of the internal stresses set up, the depth of hardening is so uniform that the stresses are adequately balanced.

Since there is little time for diffusion in this heating process it is essential that the steel be in such an initial condition that the ferrite and pearlite are uniformly distributed. An initial condition of emulsified pearlite from normalizing at a rather fast cooling rate or of sorbite from prior quenching and tempering is preferred. Steels of fine "inherent" grain size are chosen, since these have least propensity toward quenching cracks.

Mild Quenching. A method of minimizing quenching stresses that is often applicable is to avoid a steeper temperature gradient in the work than necessary, that is, to use the mildest quench that will produce the necessary hardening. Water quenching is more severe than oil quenching and air cooling less severe than oil quenching, so, if an irregular shape must be hardened that cannot be water quenched without cracking a C steel, alloy steels that can be oil hardened, or even an air-hardening steel must be substituted.

Quenching from a Low Temperature. For quenching in a given medium it is possible to flatten the temperature gradient by hardening from a lower temperature instead of from the maximum temperature to which the steel must be heated before quenching.

Ar_1 and Ar' of the cooling phase can be very materially lowered by proper alloying, even though Ac_3 of the heating phase, which has to be exceeded, may be no lower than in a C steel and to get proper diffusion the steel may have to be heated quite far above Ac_3. Once the steel has been soaked at the proper temperature, it can be cooled at as slow a rate as desired until the temperature has dropped to just above Ar'. Hence the steel can be allowed to acquire temperature uniformity just above Ar' and be quenched from this relatively low temperature. The flatter temperature gradient in such quenching introduces less internal stress.

Angell[23] has studied this matter on a wide variety of steels and in no case were the properties injured by quenching from the lower temperature, whereas cracking was greatly diminished. For example, a 0.23 C, 4.61% Ni steel was quenched (*a*) from 1450° and (*b*) from 1150° after having been first held at 1450°, and both lots tempered at 900°, the tensile, impact, and hardness results being essentially identical, in spite of the 300° difference in quenching temperature. On some highly alloyed steels it is even possible to quench from a temperature lower than that used for tempering after the quench.

Utilization of Quenching Stresses. The building up of quenching stresses is not always harmful; when the stresses can be balanced, in a symmetrical object they may do no harm; and, when the stressed portion is so positioned that the external stresses in service are applied in such a direction that they are opposed by the internal stresses, they may even be beneficial.

In making big guns by the "autofrettage" process stresses are introduced by mechanical working (through the use of internal hydraulic pressure) which oppose the pressure exerted by the explosion of the powder charge. One might consider that the steel itself is not required to withstand the initial fraction of the load of the explosion. Only after the applied pressure has balanced the stress introduced by autofrettage does the normal elastic strength of the steel come into play. Palmer[24] has used the analogy of the gun and its resistance to bursting to explain the behavior of a wire-drawing die made of carbon tool steel, internally quenched so that the bore only is hardened. The expansion during the change to martensite puts the bore in compression under a stress, Palmer finds experimentally, of 115,000 lb./in.^{2}, adding that much to the effective strength of the die. He points out that, if the die had been hardened by immersion in still water so that the outside was completely hardened and the bore not well hardened owing to steam pocketing in it, the die would be weakened for service as a die, the hole now being under tension of more than 115,000 lb./in.^{2} Thus the use of correct instead of incorrect quenching practice makes a total difference of more than 230,000 lb./in.^{2}

Palmer's whole discussion of quenching stresses is worth careful study.

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CHAPTER 7

TEMPERING AND TOUGHENING

Tempering. When a piece of C tool steel, heated to the proper temperature, is brought down through the critical ranges at a rate exceeding the "critical rate of cooling," a completely hardened steel having a martensitic structure results. Such steel is hard and brittle, and it is in a state of stress such that its use—except in a few distinctive cases—would be highly inadvisable. Under these circumstances, the application of heat will accomplish two results: it will (1) relieve the hardening strains, and (2) permit the decomposition of the martensite to proceed. By properly adjusting the temperature of this reheating process, any desired stage in the decomposition of martensite may be obtained; and, by permitting just the right amount of the hard, brittle martensite to go over into softer and tougher products, any desired combination of physical properties within the capacity of that particular steel may be realized. This process of "letting down," or softening, is called tempering.

Secondary Troostite. If the steel has been hardened fully so that it consists almost entirely * of martensite, secondary troostite will begin to form somewhere in the vicinity of 400°, or possibly lower. As the tempering temperature is progressively raised, the troostite increases in amount until at about 750° it begins to change into sorbite. Thus, steel in the tempered condition usually is characterized by the presence of more or less secondary troostite, dependent upon the degree of hardening and upon the tempering. Just as martensite may be said to represent the condition of hardened steel, or pearlite that of annealed steel, so secondary troostite is indicative of a tempered steel. The physical properties of secondary troostite are intermediate between those of martensite on the one hand and those of sorbite on the other, into which this troostite will change with further reheating to higher tempering temperatures.

If secondary troostite be tempered a bit further, it merges into sorbite, which is a non-lamellar mixture of ferrite and cementite, and not of the definite eutectoid composition like pearlite. Secondary troostite and sorbite are the most thoroughly emulsified mixtures of ferrite and cementite; they have the combination of strength and toughness sought for in quenched and tempered steels. The troostitic steels are used where hardness and strength, plus as much toughness as is compatible therewith, are of paramount importance; the sorbitic steels are used when greater toughness is needed and strength can be sacrificed. Requirements for machinability often force the use of a steel tempered to sorbite when the engineering preference would otherwise be for a troostitic steel. The hardness and toughness gradually change inversely, as tempering temperature and time increase. There is no difference in kind, only one of degree, between secondary troostite and sorbite.*

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* The common steels in the "hardened" condition consist either of maximum martensite plus austenite or maximum martensite plus primary troostite, and not martensite alone.

Secondary troostite is not nodular. It differs from sorbite only in that the cementite in the sorbite is more agglomerated. As sorbite is tempered further, the cementite particles tend to become larger and larger spheroids. The final fully spheroidized cementite structures in ferrite-cementite mixtures that can be obtained from pearlite produced at Ar_1 or primary troostite at Ar' are, therefore, the same in general make-up as those obtainable from martensite through its decomposition products, but no intermediate stage has analogous similarity.

Hardening Strains. It should always be remembered that tempering not only softens the steel through change of structure but also tends to relieve the strains set up in hardening. This last factor should not be lost from sight, for, although the proper degree of hardness is requisite for specific work, no tool will eventually prove of much value if it retains the state of strain occasioned by rapid cooling. This statement applies not only to water but also to oil, and even to air, quenching. Even the temperature of boiling water is often sufficient to relieve more or less of these stresses, if it is not desired to soften the steel further by higher reheating. Naturally, however, the higher the softening temperature the better will be the condition of the steel in this regard.

Temper Colors. Nature has provided a useful and more or less empirical indication of the degree to which tempering has affected the steel, through the formation of a surface film of oxide colors (oxide of iron). If a piece of hardened steel is brightened with emery paper or other suitable means, and then is heated slowly with exposure to the air, the brightened surface will take on characteristic "temper colors." These begin with a very faint yellow and progressively change with increase of temperature through varying degrees of yellow, brown, purple, and blue. That these colors bear a definite relation to, and are closely indicative of, a known temperature, under certain conditions, is now a generally accepted fact. Although a difference in distinguishing the various shades of color is bound to occur because of the "personal equation," the following table is fairly representative:

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* The stress-relieving, low-temperature treatment and the treatment for formation of secondary troostite are commonly termed "tempering," whereas higher temperature treatment that "draws the temper" and appreciably softens the steel is termed "drawing." However, these processes, like their products, differ only in degree, and the reheating of a hardened steel is rather indiscriminately termed either tempering or drawing.

Temperature, Deg.	Color	Temperature, Deg.	Color
420	Very faint yellow	510	Brown
430	Yellowish-white or light straw	520	Brown-purple (peacock)
		530	Light purple
440	Light yellow	540	Purple
450	Pale yellow straw	550	Dark purple
460	Straw	560	Light blue
470	Dark yellow	570	Blue
480	Deep straw	600	Dark blue
490	Yellow-brown	625	Blue tinged with green
500	Brown-yellow		

Colored plates illustrating these shades are given in accessible publications.[1, 2]

Limitation of Color Method. The previous statement regarding the relation of temper colors to temperature is true in its entirety only under certain definite conditions of heating, which are dependent largely upon the time element. So long as the temperature of the steel is being raised progressively, that is, so long as the temperature of the fire, furnace, or tempering plate is greater than the temperature of the steel, the temper colors indicate the temperature of that part of the steel most affected—the surface. But when the steel is kept at a definite tempering temperature for any length of time, the colors do not represent the actual temperature. This point is readily illustrated by heating a small piece of hardened steel at a constant temperature for a considerable period of time. Thus, in one instance, a straw color was produced in about a minute, but changed to a brown in about 10 min., and to a purple in about 40 min.; and yet the temperature of the steel was never higher than 460°, representative of the straw color. In other words, the time element has developed a new set of conditions which may greatly affect the depth of oxidation, or color.

On the other hand, it is a debatable point whether or not these temper colors represent the actual condition (not the temperature) of the steel itself. Some tool makers maintain that the efficiency of the tool, both in hardness and in other properties, is the same whether the color has been obtained by a short heating at a high temperature or a longer heating at a lower temperature. That is, the ultimate results are indicated by the temper color, independent of the method of obtaining it. Others aver that such is not the case.

Tempering for Depth. It is obvious that the temper color is at the best but a surface indication. For some tools or articles which require a specific superficial hardness only, and in which the condition of the center of the tool is of little consequence, it probably does not matter a great deal in the ultimate results whether the temper color—a straw color, for example—has been obtained by a few minutes' heating at 460°, or by heating for a longer period at, say, 400°. Contrariwise, if the tool or part is to be subjected to stresses of such a nature as to demand the best that the steel is capable of, the greatest degree of uniformity and release of hardening strains are required. Such may be obtained only by a thorough heating at a specified temperature, which may be entirely independent of the color indication. In such cases the thorough heating at 400°, which affects more uniformly the whole mass of the steel, might prove immeasurably better than the incidental surface heating to 460°. As will be mentioned later, a continued heating at 460° would be an improvement over either color method.

Quenching after Tempering. The method of tempering by color indication inherently requires rapid cooling when the specified color is reached, to prevent any further rise in temperature, or, in the blacksmith's phrase, to "set the grain." Although it is possible to heat the steel so carefully that the maximum effect is just to develop the color desired and to carry the heating no further, such methods take so much time and patience that they rarely are carried out in practice. The necessity of cooling by quenching, even if the quenching is done by an experienced hardener, is the source of many troubles. Not only does the quenching probably induce further strains into the steel, but it is also entirely inconsistent with uniformity of results. If the object is of considerable size, or varies greatly in dimension of adjoining sections which are to be similarly tempered, or is of intricate design, the difficulty experienced in obtaining the same temper throughout, even on the surface (to say nothing of the interior of the steel), will be greatly magnified. If the proper color is reached on one part before another, there will be a corresponding difference in hardness. Thus the difficulties multiply.

Use of Liquid Baths. Later methods involving the use of liquid baths for heating overcome the difficulties in color tempering, eliminate, as a general rule, the necessity for quenching, and, furthermore, give complete uniformity of heating throughout the whole mass of the steel and the maximum elimination of hardening strains obtainable at the temperature used. If the bath is maintained at the proper temperature there can be no overheating, the heat must penetrate all parts of the steel alike, and the "personal equation" is as nearly eliminated as possible. This method has the further advantage of cutting down labor costs and increasing the output, since a number of pieces may be heated at the same time, and while one lot is being tempered another bath may be charged or discharged.

Comparison of Physical Properties Obtained. An excellent example of the efficiency of bath tempering is illustrated in automobile gears. Because of the relatively thin section of the teeth as compared with the mass of the gear, exact tempering by the ordinary temper-color practice is rather difficult. The teeth, which should be the hardest, take the temper first, and are therefore the softest part of the gear as a whole. If the gears were to be tempered by revolving on a hot bar, much better results would be obtained than by ordinary tempering, but the time and cost would prove excessive where hundreds of pieces were to be handled. By the use of a suitable liquid tempering bath, thorough practical uniformity can be obtained throughout. Whereas, by the color method, the core of the gear would have the tendency to be too hard, the teeth perhaps too brittle or soft in places, and only the surface of the gear as a whole affected by the temper color representing, say, 475°, by the more modern method the whole mass of the gear would have the physical properties characterizing the drawing temperature of 475°.

Exact Temperatures. Too much attention cannot be given to the necessity of obtaining exact temperatures in the tempering operation. For the average run of C tools the tempering range is very narrow, probably within 100° for the great majority. The tempering action takes place extremely rapidly and often a difference of 15 or 20° may cause much trouble. Trying to temper tools over an open fire may be permissible in isolated instances, but it spells failure if made a general practice.

Tempering Methods. The procedure to be employed in tempering must necessarily depend upon the nature of the tool or part. Methods must be developed to satisfy the individual requirements and are too numerous to discuss here. Briefly, however, the more common practices may be covered by the tempering plate, the sand bath, and such liquid baths as oil, lead and alloys, and molten salts.

Tempering Plate. The tempering plate generally consists of an iron casting planed smooth on top, and heated from beneath by suitable means. The steel articles are placed on the plate, moved about until they have attained the proper temper color, and then quenched.

Sand Bath. In order to effect more uniform tempering of small tools, a pan of clean, well-dried sand may be placed on a suitable hot plate, or in a furnace. The sand is held at the desired temperature, which may be determined by the insertion of a thermometer or pyrometer couple, and may be protected by covering with a suitable hood. The oxide colors on the steel may also be used as a measure of the tempering, as there is, of course, free access of air between the particles of the sand.

Oil Baths. For much of the ordinary tempering work, an oil bath probably will prove as satisfactory as any method for temperatures up to about 500°. The chief requisites are a tank holding an ample supply of oil, a suitable furnace or method of heating by which accurate and constant temperatures may be obtained, and a means for determining the temperature of the oil. Mineral oil with a flash point of some 600° is generally used for the bath; certain of the animal and vegetable oils are also used occasionally.

Handling the Material. The oil baths are provided with a wire basket in which the pieces to be tempered are placed and which then is lowered into the oil. The salt baths are similarly equipped. By this method a number of pieces may be tempered at once, and the steel is prevented from coming in contact with the sides or bottom of the tank, which is likely to be hotter than the oil. It is advisable, whenever possible, to allow the hardened steel to come gradually to the desired temperature, and not to immerse it in the oil when the latter is already at the highest heat. It is better to put the steel in the oil when the oil is about 200 to 300° and let the two heat up together. The reason for this is that the preheating—if it may be so termed—allows the heat to penetrate more gradually, softening the outer portion of the steel in such a way that the inner and stressed part may be relieved more gradually, thus avoiding the danger of fracture. Sudden heating has a tendency to set up new stresses which in turn must be overcome. The length of time allowed for the tempering will depend upon the size and nature of the piece under treatment; 15 min. or so, after the maximum temperature has been reached, generally will be sufficient for the average run of small tools, gears, etc., whereas larger parts require more time in proportion. If large and small parts are tempered at the same time, it will do no harm to the small pieces if they are not removed until the larger pieces are ready. When the full effect of the tempering has been attained, the pieces may be removed from the oil and allowed to cool in the air, for if the steel has been heated thoroughly at the maximum temperature of the tempering operation, no further change will take place in ordinary steels. Each phase of the transition is represented by a definite temperature for each steel, so that no further step in the transition will occur unless the temperature is raised—with the possible theoretic exception of very long-continued heating. For some large work, such as die blocks, large cutters, etc., the steel is allowed to cool in the oil in order to procure the greatest possible elimination of strains.

Salt Baths. If higher drawing temperatures than those possible with oil are desired, a bath of salts may be used. A combination of two parts of potassium nitrate and three parts of sodium nitrate melts at about 450° and may be used up to about 1000°. The use of nitrate salts instead of the chloride salts is necessary because of the lower temperature desired. (See Chap. 13.)

Lead Baths; Alloys. Lead, having a melting point of 620°, may be used for tempering where temperatures higher than its melting point are required. The melting point may be lowered by alloying the lead with tin. Thirty per cent tin gives a bath completely fluid at about 500°, and 60%, a bath fluid at 400°.

Forced Convection Furnaces. For use at temperatures so low that oxidation by the air is inappreciable, there are modern tempering furnaces, usually heated electrically, in which hot air is circulated through the charge by a fan, thus bringing a steady supply of heat to each article even in a rather closely packed mass. By using a neutral, nonoxidizing gas recirculated through a heater, the method can be applied up to quite high temperatures. Such schemes provide much improved uniformity of heating over those of heating in air or gas without circulation. Such furnaces have largely displaced the older methods. They will be further discussed in Chap. 15.

TOUGHENING

Sorbite. As the reheating or drawing temperature is increased still further beyond the tempering range, another stage in the decomposition begins—the change of troostite into sorbite. Like the change from martensite to troostite, the formation of sorbite does not take place spontaneously throughout the steel, but increases gradually and progressively. Sorbite tends to coagulate into spheroidized cementite and ferrite. On higher heating, sorbite becomes slightly coarser, then slowly becomes definitely coarser, and finally tends toward the spheroidized state. Sorbite differs from troostite in that it is softer for a given C content and etches without smudging.

In general, secondary troostite or tempered martensite has carbide particles which are so fine that they are unstable to etching reagents and therefore decompose (black smear and quick darkening), whereas the larger carbides (produced by higher or longer tempering or both) can be retained and seen as such in polished and etched specimens.

Importance of Sorbite. The importance of sorbite is due mainly to its physical properties. Although sorbitic steel is slightly less ductile than pearlitic steel for a given C content, its tensile and yield strengths are so high that a higher combination of these three properties can be had in sorbitic than in pearlitic steels. Steels that are so treated as to contain sorbite are often called "toughened" steels.

Toughening Range. The transition of secondary troostite, the chief characteristic of tempered steel, into sorbite, characteristic of toughened steel, is gradual; it progresses with the increase and duration of the reheating. At some point, depending upon the composition of the steel and the degree to which it has been affected by the hardening process, sorbite is formed. With sorbite as the characteristic constituent of toughened steel, we may consider as the lower limit of the toughening range that temperature which will produce sorbite. In fully hardened steel of the medium forging and higher-C analyses, characteristic sorbite begins to form at about 750°. At about 1250 to 1300°, under suitable conditions of time, the sorbite tends to coagulate, and takes on the spheroidized structure previously mentioned. (See Fig. 1.)

Thus the toughening range lies approximately between 750 and 1250°. It must be remembered, nevertheless, that these temperatures are in no sense definite, but are taken arbitrarily as representative of a class of heat-treatment work: differences in chemical composition, the degree of hardening, the size of work, etc., all play their part.

Influence of Toughening. When a piece of hardened steel is reheated for toughening, each specific temperature has a certain definite influence upon the steel. The results of this toughening process are interpreted by the ability of the steel to do certain work, to withstand the application of stated loads, or are measured by standard methods of testing.

Effect of Increased Temperature. Each increase in the toughening temperature lowers the tensile and yield strengths, with a corresponding increase in the ductility. With the majority of ordinary C common-alloy steels, the ratio of the yield strength to the tensile strength remains very nearly constant throughout the sorbitic range (approximately 750 to about 1250°). Beyond these temperatures, and coincident with spheroidizing, and, later, with annealing and the formation of pearlite, the values for the yield strength and tensile strength —of each particular steel—begin noticeably to diverge until they reach their smallest ratio in fully annealed or spheroidized steel. Almost to the end of the sorbitic range, the graphs obtained by plotting the yield strength and tensile strength against the drawing temperatures are, for general purposes, straight lines; but beyond this range the curve tends toward the horizontal. (Compare Fig. 19, Chap. 1.)

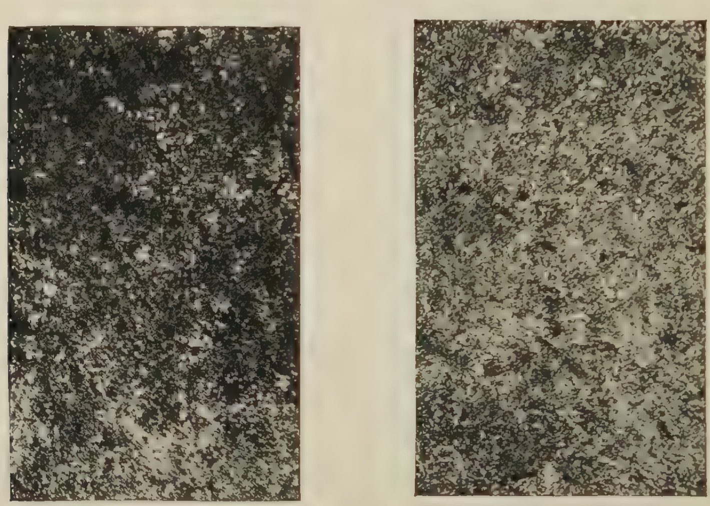
(*a*) 800° draw. (*b*) 1250° draw.
× 100 (Bullens.)

Fig. 1.—Sorbite, illustrating the effect of increase in draw temperature. Coagulation has begun in (*b*).

Effect on Ductility. These changes are accompanied by reverse changes in the ductility, as measured by the reduction of area and elongation. These two measures of ductility differ in that the reduction of area generally reaches a maximum at about the end of the sorbitic and spheroidizing range and then decreases, whereas the elongation does not attain its maximum until the steel is fully annealed. However, through the sorbitic stage at least, each increment of decrease in tensile strength and yield strength is associated with, and counterbalanced by, an increase in the reduction of area and elongation. The change in static strength and ductility are, furthermore, almost directly proportional to the toughening temperature.

Effect of Long Time Tempering. Although it is possible to obtain the tensile and yield strengths, and elongation indicative of a definite toughening temperature by means of a minimum heating at that temperature, the reduction of area is often remarkably increased by prolonged saturation. This, and the practical benefits obtained from it, have been brought out very clearly by the experience of rifle-barrel manufacturers: barrels (1$\frac{1}{32}$ in. diameter) that had been heated for one-half to one hour at the drawing temperature would pass the physical test, including a reasonable reduction of area; but when it came to the machining, and particularly the drilling and rifling, great difficulty was experienced and many rejections resulted. Barrels that had been "soaked" at the same temperature for two to three hours showed approximately the same tensile and yield strengths, and elongation as with the shorter saturation, but gave, in addition, a large increase in the reduction of area and superior drilling qualities. Prolonged tempering was the secret of the heat-treatment process in this case, and many other examples might be given to prove its tremendous advantage.

Capacity of the Steel. Thus it will be seen that by changing the drawing temperature the grouping of these factors may be varied through a considerable range and may be limited only by what we may call, for want of a better phrase, the "capacity of the steel." This quantity is defined largely by the chemical composition, the method of manufacture, the size of the piece to be treated, and other subordinate factors. With these qualifying conditions in mind, we may further define the capacity of the steel as the limiting ratio of strength to ductility. Each steel, as qualified above, has certain definite limits within which the physical properties may be varied. At one end of the see-saw, as in hardened steel, there is a maximum tensile strength with minimum ductility; and at the other extreme, as in fully annealed, coarsely pearlitic steel, there is a minimum tensile strength with maximum ductility. Following out the simile of the see-saw, we may place tensility on one end and ductility on the other; when one is up, the other must be down; both cannot be up or both down at one and the same time; raise one and the other must fall. The heat treater now stands on the middle of the board and by means of his choice of annealing, normalizing, or quenching and tempering, and especially by his choice of reheating temperature in tempering can adjust the opposing factors to that position which he desires; but he cannot change the maximum and minimum of either, because they are fixed by the limitations previously mentioned at the beginning of the paragraph, and over these he has no control as far as the individual steel is concerned.

Duplication of Results. Happily for the operator, each grouping is distinctive of a definite toughening temperature, other conditions being the same. When he has once determined the relation existing between static strength, ductility, and temperature, for a given sized piece of work made from a steel of specific analysis and grain size, he knows that he can approximately duplicate his results under like conditions at any time. Not that he can absolutely and ultrascientifically obtain results within a few pounds yield strength or hundredths of a per cent elongation—for such control is neither necessary nor expected—but that he can reasonably expect to get a commercially acceptable duplication.

Slow Cooling and Stresses. Slow cooling predicates the release of internal stresses. Not only is this true of the full-annealing process, slow cooling from a temperature above that of the critical range, but also of the toughening operation. In fact, the very nature of the usefulness of toughened steel depends upon the absence of a state of stress just as much as upon specific static or dynamic properties. Some of the failures in locomotive forgings may be traced back to the lack of slow cooling after toughening; and this trouble sometimes justifies the requirement of cooling in the furnace especially on very complicated objects. Just as the dangers in hardening increase with the rapidity of cooling, carbon content, and size of section, so are they likewise magnified in cooling after toughening, although on a smaller scale. If these factors become noticeably important, cooling in air from the toughening temperature may set up such a new series of cooling stresses that many of the real advantages of toughening may be invalidated.

Use of Furnace Cooling. The greater part of hardened and toughened work, such as automobile and other small forgings, may not require furnace cooling, apart from the fact that it may be economically impracticable. But even with such work it is desirable that the pieces be piled together after removal from the furnace so that the cooling will be retarded. For forgings of thickness greater than 3 to 4 in., such as heavy machine parts, ordnance, etc., cooling in the furnace is always desirable. It may be said that such cooling never did any harm, in so far as C steels are concerned, and it may do a world of good in relieving strains.

The important exception to slow cooling from the toughening temperature is that of certain CrNi, Mn, and other similar alloy steels. Under distinctive conditions for these steels, rapid cooling—even quenching—is required to eliminate what is known as "temper brittleness," or "Krupp Krankheit," in relation to notched-bar values. This will be discussed more fully in Chap. 9, Vol. II.

High *vs.* Low Toughening Temperatures. On the hypothesis that either of two specified analyses would prove equally satisfactory, under suitable treatment, for the same piece of work, but that on account of the difference in chemical composition one steel would require toughening at, say, 1200° and the other at, say 800 or 900°, the selection of the higher drawing-point steel should be made. Such conditions often arise in heat-treatment plants handling a variety of commercial work, and it may be well to sum up briefly the reasons for the above conclusion.

The more stable approach to equilibrium which exists between the transition constituents, the more lasting and effectual will be the treatment. Furthermore, the smaller the amount of internal stress which may remain in the steel from the previous hardening operation the better. Both these conditions are more nearly brought about by the higher drawing temperature.

From the furnace man's point of view, the temperatures around 1200°, being of characteristic visible reds, are recognizable decidedly more easily than those around 800 to 1000°, since with these lower temperatures there is very little visible heat color. The higher drawing temperatures, therefore, aid in the efficiency of judging the heating operation and lead to greater uniformity of control and of results, when they have to be carried out without pyrometric control.

Quenching Medium *vs.* Toughening Temperature. There is another phase of the high or low toughening temperature proposition which cannot be solved by any general rule, but only after due consideration of all the circumstances involved; this relates to the condition of affairs in which there is no opportunity for the choice of steel, and the success of the operation depends more upon the selection of the quenching medium in relation to the toughening temperature. As we have noted, water quenching gives a harder steel than oil quenching. It naturally follows that, in order to obtain approximately the same physical results, the oil-quenched piece must be drawn at a lower temperature than the water-quenched piece. The arguments regarding water *vs.* oil quenching, and low *vs.* high drawing temperatures have been discussed previously. If the solution were to be developed entirely along these lines it is probable that in the majority of cases the oil quenching (giving less hardening strains) and lower drawing temperature would be employed. In other words, the difficulties to be encountered with water quenching—the hardening operation being the more drastic of the two—would more than outweigh the disadvantages of the lower toughening temperature. This is a question in which the personal element and experience of the operator would be paramount.

Influence of the Carbon Content. In respect to the selection of the steel in relation to the treatment, there remains the consideration of the influence of the C content. Carbon not only intensifies the effect of the rapid cooling (hardening), but also directly augments the brittleness of the steel. Or, to put it in other words, the greater the C content the greater the hardening stresses, and the lower the ductility which can be obtained with a stated tensile strength. It usually is desirable, therefore, to provide a steel with as low a C content as will give the desired results. This might be repeated with even greater emphasis in relation to alloy steels. As alloys are introduced, the C content must be correspondingly reduced if toughness is to be maintained.

Toughening *vs.* Annealing. It is only within the last few decades that the toughening process with its attendant sorbitic structure has been used and understood. Previously, annealing was generally the cure-all for brittleness and a stressed condition of the steel. Pearlite—produced by annealing—because of its entangled structure, gives a large measure of ductility; but also gives a minimum tenacity. The structure of sorbite, however, is even more entangled than that of pearlite; sorbite is far superior to pearlite in tensile strength and especially in elastic limit. Thus, by obtaining a sorbitic steel by suitable treatment, it is possible to secure almost as much ductility, greater strength, and—by the use of a lower-C steel—less brittleness than in a pearlitic or annealed steel.

Standardization of Results. With the same degree of hardening, if the reheating has been uniform and thorough at a given temperature, the physical results will be comparatively the same for material of equivalent section and the same composition. That is, the product will be standard for standard treatment. Furthermore, in order to get standard results with steel purchased under the same general specifications (i.e., each chemical constituent within certain limits), the toughening temperatures may be varied according to the chemical composition. To illustrate: the following heats of steel of varying chemical composition and made by several steel companies were manufactured into a certain product which, when heat treated, required a yield of 85,000 to 95,000 and an elongation of not less than 16%. In spite of the varying C, Mn, Cr, and Ni contents, the toughening temperatures (maintained within ± 5°) were so adjusted as to give the desired results.

C %	Mn %	P %	S %	Cr %	Ni %	Toughening Temp.
0.16	0.43	0.015	0.017	0.62	1.82	1050
.185	.44	.010	.015	.57	1.56	975
.20	.43	.009	.016	.64	1.74	1025
.20	.46	.011	.017	.40	1.56	950
.21	.48	.015	.015	.60	1.77	1050
.21	.50	.017	.014	.67	1.84	1075
.23	.50	.016	.018	.65	1.73	1075
.245	.53	.015	.020	.62	1.79	1120
.25	.50	.011	.019	.64	1.40	1140
.26	.43	.010	.018	.60	1.65	1100
.27	.49	.015	.021	.63	1.79	1150
.28	.51	.008	.011	.41	1.57	1120

Thousands of pieces, some weighing as much as 200 lb., fulfilled, by actual test, the standard specifications.

Quench Toughening. A process which has been used considerably for the treatment of large forgings of uniform section, such as heavy axles, is heating as usual for hardening and then quenching for a specified number of seconds, followed by air cooling. The quenching affects the steel to a certain depth, but still leaves a considerable amount of heat in the forging when removed from the bath. As the forging cools in the air this heat from within will toughen or "soften" the steel affected by the quenching. In order to obtain equivalent results under varying conditions, the number of seconds required for immersion in the oil of a piece of given size must be determined by experiment and strictly adhered to. Forgings treated by this process are characterized by a soft or annealed core, with a progressively toughened outer part.

Rail ends are heat treated [3] by spraying a regulated amount of water on the hot rail and letting the heat in the rail body temper the quenched surface layer. Whole rails have been heat treated by timed immersion in water, then withdrawing for tempering, in analogous fashion. Epstein [4] discussed somewhat analogous methods suggested by others. Such methods are difficult to control and will be avoided where possible.

Tempering after Cold Work. Tempering after cold work may be carried on at so low a temperature that no appreciable change in microstructure occurs, but there is a release of internal strain. By cold working, the tensile strength is raised and the ductility diminished.

The stress-strain diagram of a cold-worked metal shows a low proportional limit and a gradual curvature beyond it. By low temperature tempering the curve is altered, there being a longer range of elastic action and a more abrupt transition into plastic action. Thus the load-carrying ability without undue deformation, i.e., the elastic limit or yield strength, is increased whereas the increased tensile strength is not lost, and the ductility and impact resistance are but slightly decreased.

Mann[5] shows a diagram, Fig. 2, to illustrate this. Slow recovery from overstrain may occur at room temperature, softer steel recovering more rapidly than harder ones. By low temperature treatment, with the temperature chosen to suit the steel, but so low that no visible microstructural softening occurs, the recovery can be accelerated and the steel put in stable condition.

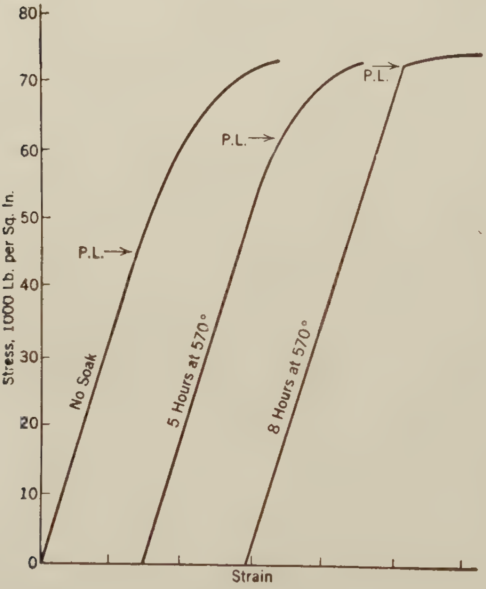

Fig. 2.—Effect of tempering on the stress-strain diagram of cold-worked steel. (Mann.[5])

Removal of Residual Stress by Tempering. The tempering temperature after cold working for various steels will run from 400° on soft steels to 750° or a little above on the more highly alloyed steels. Temperatures of 475° for 0.22% C steel and 750° for a 0.26 C, 2.50 Ni, 0.55 Cr, 0.52% Mo steel were found suitable by Macrae.[7]

Tempering cold drawn screw stock at 750° is considered to improve its machinability. Relief of residual stress after cold work has been experimentally studied by Stewart[8] for steels of 0.27 C, 0.43 Mn, 0.01% Si and 0.21 C, 1.44 Mn, 0.01% Si. The latter showed higher residual stress. Process annealing above 800° is required to produce plasticity enough to start much relief of stress even in 4 hr., and a temperature of 1200° for 1 hr. is considered necessary to reduce residual stress to a negligible figure. The effect of welding procedure on residual stress is also discussed.

Autofrettage. This phenomenon is utilized in strengthening big guns and the like by stretching them radially and longitudinally by internally applied hydraulic pressure so that the bore is slightly expanded and the steel thus cold worked. When the internal pressure is released the gun is left with compressive stress in the outer layer, and the gun is strengthened much as when an outer gun tube is shrunk upon an inner tube or when the gun is tightly wound with wire. This autofrettage process is widely used at arsenals, including those of the United States. The low temperature treatment is essential in giving the gun the yield strength desired against the internal pressure of the charge.

The method is applied both to forgings and to centrifugal castings. Col. Dickson[6] has described the process as applied at Watertown Arsenal to centrifugally cast guns. In one case a 0.35 C, 0.60 Mn, 0.30 Mo, 0.10% V steel, after normalizing from 1740° to break up the cast structure, was full annealed by furnace cooling from 1360°, then water quenched (twice) from 1360° and drawn at 1290°, after which it showed 96,000 tensile, 77,000 prop. limit, 24% Elong., 60% R.A., 35 Charpy on tangential specimens.

It was then cold stretched by enlarging the bore some 40% under internal pressure of 105,000 lb./in.^{2} and soaked at 570°, when the properties were:

114,000 tensile, 103,000 prop. limit, 12% Elong., 51% R.A., 27 Charpy.

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CHAPTER 8

PRECIPITATION HARDENING

It is often considered that low-C steels are not affected by quenching and tempering, which is true in the sense that such steels, as has been shown in Chap. 6, demand an extremely high rate of cooling to produce martensite, so that unless the steel is in very small section and the quenching very drastic, an attempt to quench results only in the usual normalized, rather than the hardened structure.

Behavior of Low Carbon Steel. If, however, a very low C steel, say 0.05% C, is quenched, not from above Ac_3, say 1700°, as usual, but from below it, about 1300° and then tempered for perhaps 10 hr. at 140°, it will be found that the Brinell hardness as quenched may be 120 and as tempered 150. If we let the steel merely stand at room temperature, it gets harder, in a month it will reach 170 Brinell, and its elongation may drop from an original 25% down to 15%. A similar but less marked change is noted in higher C steels, quite discernible at around 0.15% C; but in still higher C steels the phenomenon is less readily observed because of the higher hardness of such steels.

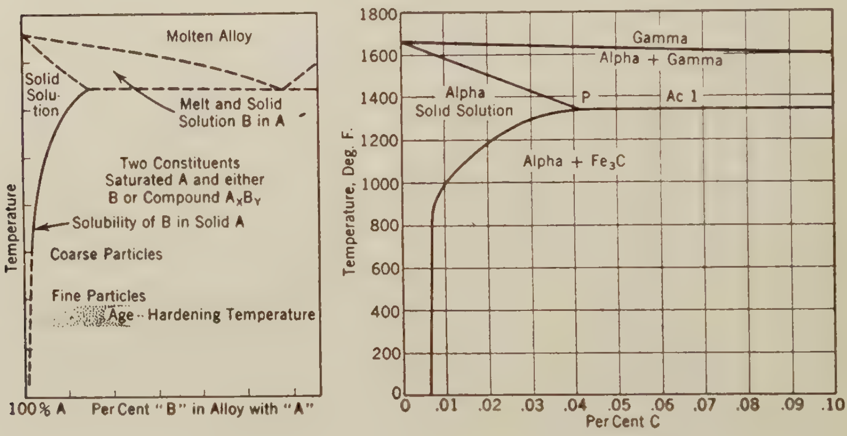
Briggs,[1] however, found that an ordinary cast steel of 0.25 C, 0.34 Si, 0.70% Mn quenched from 1250° and aged for 2 hr. at 212°, 20 hr. at 120°, or allowed to rest at room temperature for 30 days, increased in Brinell hardness from 155-160 annealed, and 160-165 as quenched to 175 for the elevated temperature aging and 190 for the room temperature aging. The aging increases tensile and yield strengths with decrease in ductility. The properties of this steel annealed at 1650° with and without subsequent water quenching from 1250° and 20 hr. reheating at 120° were:

	Annealed Only	Annealed, Quenched, and Aged
Tensile	79,000	85,500
Yield	55,000	63,000
Elong. %	25.5	19.5
R.A. %	40	33

The hardness of the material aged at 120° did not change after 6 months storage at room temperature. In discussion, Herty stated that rolled steel of the same composition did not age harden under the same treatment and ascribed the difference to the microsegregation of C in the grains of the cast steel as against more uniform distribution of C in the wrought steel.

Briggs' tests were made on 0.505-in. diam. bars. The indications are that a rapid rate of cooling is necessary in the quench to put the steel in condition for precipitation hardening, so that the treatment would probably not be effective on large castings with heavy sections.

Mechanism of Precipitation Hardening. From such observations as those above it is deduced that the effect is due to solution of cementite in ferrite at 1200-1300° and precipitation of cementite from ferrite at lower temperatures. Further study shows that the solubility curve for cementite in ferrite has the shape shown in Fig. 1*b*.[2] Its absolute location is still somewhat in doubt.



(*a*) (*b*)

FIG. 1.—(*a*) Typical constitution diagram illustrating the solid solubility characteristics necessary for age-hardening. Schematic diagram. (Davenport and Bain.[2]) (*b*) Solubility of cementite in ferrite. This is an enlarged plot of the same field in Fig. 7, Chap. 2.

This has the form of the typical "precipitation hardening" curves, schematically shown in Fig. 1*a*, and familiar to nonferrous metallurgists, especially in respect to duralumin and to copper beryllium alloys. The behavior of the quenched material on tempering at different temperatures and times is like that of precipitation hardening alloys in general, the hardening taking place so slowly that the process is usually called "aging" or age hardening rather than tempering. The term tempering, which usually connotes some degree of softening, is not suitable, because in these cases tempering does not progressively soften the material.

The results of quenching and reheating in a precipitation-hardening alloy are just reversed from those of martensitic quenching. In the case of ferrite and cementite, like other precipitation-hardening alloys, quenching *softens* and *toughens,* whereas reheating to a lower temperature *hardens.*

However, long continued reheating will again soften the material, as Fig. 2 shows.

Solution and Separation. The explanation for these phenomena is that at the quenching temperature, above the solubility line, cementite is taken into solution and retained in solution on quenching. In the cold state the ferrite cannot firmly hold the cementite in solution; the cementite wants to precipitate out. If the steel is warmed somewhat, or even on very long standing, the cementite does come out, in very tiny particles. At room temperature these tiny particles remain tiny and distributed, but at slightly higher temperatures they diffuse, agglomerate, and spheroidize, just like the more notable spheroidization in a spheroidizing anneal of pearlite. In the agglomerated condition these larger particles of cementite do not have as much hardening effect as when they are smaller and more uniformly distributed. This effect is shown in Fig. 2.

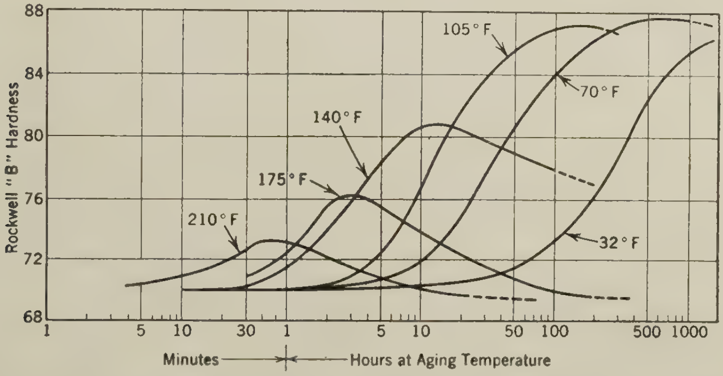


FIG. 2.—Quench aging of 0.06% C steel after quenching from 1325°, plotted on logarithmic time scale. (Davenport and Bain.[2])

Such phenomena occur within the ferrite of ordinary steels on cooling from temperatures above the cementite solubility line. If the steel is not quenched from above the solubility line, but slow cooled, the carbide precipitation may occur in grain boundaries and cleavage planes and embrittle the steel, whereas, if it is quenched and then allowed to stand, or only slightly reheated, the carbide will precipitate throughout the grains and embrittlement will not ensue, merely a hardening and strengthening will occur.

Temper Brittleness. This "quench-aging" phenomena is not very much of a help or very much of a nuisance in ordinary C steels. In certain alloy steels, notably those containing considerable Mn, say 1.25-2%, and those with commercial amounts of Ni and Cr, it is found that after the regular martensitic quenching and tempering, with tempering temperatures around 800-1000°, for instance, the steels are brittle in impact if slowly cooled from the tempering temperature, but tough if quenched from that temperature. This is termed "temper brittleness" or, in German, "Krupp Krankheit," i.e., "Krupp sickness," since the Ni-Cr steels in which it is prevalent were early known as Krupp compositions.

Cementite, of course, is not pure Fe_3C in steels alloyed with carbide-forming elements, like Mn, Cr, Mo, V, etc., but instead is a more complex carbide. The complex carbides containing considerable Mn or Cr seem peculiarly prone to grain boundary precipitation on slow cooling of ferrite in which they are dissolved, and of course the presence of other metals dissolved in ferrite affects the ability of the ferrite to dissolve and hold or to precipitate out the carbides.

Although the exact solubility curves for carbide in these temper-brittle alloy steels are not known, it is plain that a precipitation hardening effect, and almost certainly one connected with precipitation of carbides, is involved.

This is a commercially important matter, for whereas temper-brittle steels show normal ductility in the tension test, they are poor in impact, and considered undesirable for shock-resisting service. The magnitude of this effect is rather large. A temper brittle 0.30 C, 3.50 Ni, 1% Cr steel may show 5 ft. lb. Izod on furnace cooling after tempering and 50 if water quenched from the tempering temperature.

Not all heats of compositions that are normally temper brittle turn out to be temper brittle, and the range of tempering temperatures that produce temper brittleness may differ from heat to heat, but rapid cooling after tempering is a wise precaution when handling this class of steels. If the section treated is so large that the interior cannot be cooled rapidly by quenching, an alloying element should be added that will minimize temper brittleness. Fortunately Mo has this effect. If a part of the Mn or Cr of temper-brittle steels is replaced by, say, 0.20% Mo, the steels are free from temper brittleness. Cementite carrying Mo appears to diffuse less readily and not to be precipitated at the grain boundaries.

Size of Particles. In all this discussion of precipitated carbides and other precipitation-hardening materials it should not be thought that the carbides or other precipitated materials are necessarily visible under the microscope. As a rule they are not visible till they have been agglomerated and spheroidized to such a size that the matrix ceases to be appreciably hardened by their presence, but the existence of submicroscopic hardening particles is thoroughly accepted by metallurgists on the basis of their effects, even when invisible.

Strain Aging. In precipitation hardening in general, it is found that cast material responds less readily to solution and aging treatments than does work hardened, i.e., strained and deformed material.* This effect crops up in cold-rolled, low-C steels. These rimmed, unkilled steels, as usually made, contain both C and O, as well as small amounts of N. Bessemer steels are especially high in N.

Now O and N, or more strictly, Fe and Mn oxides and Fe nitride show solubility curves in ferrite of the same general character as that of Fig. 1 for cementite in ferrite.†

In quench aging and temper brittleness, it is quite certain that the carbides are primarily responsible. But aging after cold rolling, i.e., strain aging, appears to be more definitely due to precipitation of oxides and nitrides than to carbides, though all may be concerned.

Separation During Testing. The assumed situation in strain aging is that the steel has in solution, not previously precipitated, some compound that wants to come out. During deformation, as in cold rolling or in compression or tensile testing, it starts to come out, and it may continue to be precipitated thereafter at room temperature, or more quickly on slight reheating. Such compounds so precipitated tend to come out in the grain boundaries and cleavage planes and make the steel more brittle. The impact resistance of cold-worked and aged steels, compressed and allowed to age, is shown in Fig. 3 from Epstein.[3]

The precipitation effect is so great in ordinary mild steel that it occurs while a tensile bar is being pulled at slightly elevated temperatures, as is shown by the peak in the tensile strength of the mild steel in Fig. 4.[4]

Cold-worked, low-C steel tested immediately after being slightly

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* An exception noted in the work of Briggs and of Herty on quench aging of cast and wrought 0.25% C steel, noted at the beginning of this chapter, does not affect the validity of the general rule.

† The solubility curve for Fe nitride, according to Séferian,[6] shows dissimilarity at a N content above that present in the steels under discussion.

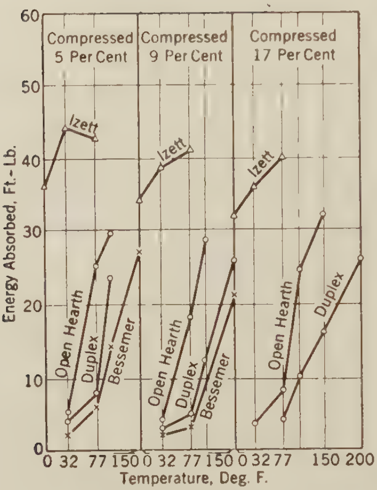

FIG. 3.

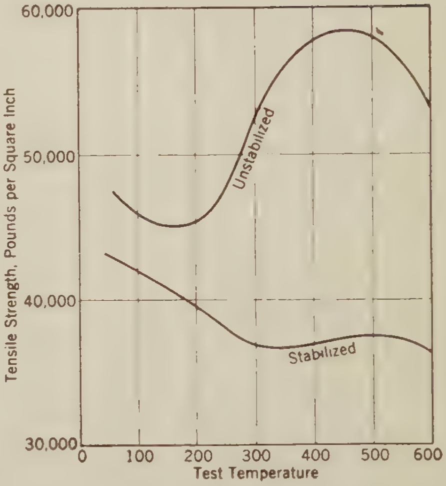

FIG. 4.

FIG. 3.—Effect of lowered temperature on impact resistance (Charpy bar, keyhole notch) after various degrees of cold work by compression. The Izett steel was treated with Al, which is considered to combine with O and N to form insoluble compounds, while the other steels contain soluble ones. The sensitivity to lowered temperature increases with the amount of soluble O and N. (Epstein.[3]) FIG. 4.—Tensile strength at elevated temperature of ordinary low-C steel (unstabilized) and of similar steel stabilized by addition of Al and Ti. (Kenyon and Burns.[4])

cold worked has a smooth tensile stress strain diagram.[5] If allowed to age it shows a jogged one, Fig. 5.

Stretcher Straining. The presence of this jog leads to a difficulty in deep drawing known as "stretcher straining." Davenport and Bain [2] have cited evidence which shows that stretcher straining is due to a precipitation-hardening action.

Stabilization. Moreover, as Griffis [5] and co-workers have shown, addition of 0.04% Al + 0.06% Ti to a 0.05% C steel stabilizes the steel and prevents the action, whereas, as Epstein [3] has brought out, the Al addition to "Izett" steel also prevents strain aging. Since Al and Ti can readily combine with O and N to form inert oxides and nitrides that will not readily go in and out of solution, this effect makes it pretty clear that in the "unstabilized" steels, it is a precipitation-hardening effect that is concerned.

Utility of Precipitation Hardening. These cases of precipitation hardening produce phenomena that one generally seeks to avoid, and they are commercially advantageous only in rather special cases. Carbide precipitation effects are of great moment in some of the

Cr steels that are welded or that are employed in high temperature service. Not only the ferritic steels, but also the austenitic ones, like high-Mn steel and the familiar 18 Cr 8% Ni stainless steel, are subject to them. Such phenomena will be mentioned in Chap. 20, Vol. II. However, there are cases where precipitation hardening can be usefully applied to an alloy steel. There are many highly interesting cases of precipitation-hardenable ferrous alloys, most of which are too highly alloyed to come into the scope of this book. One case, involving small amounts of a cheap alloying element, Cu, is of interest to the general heat treater.

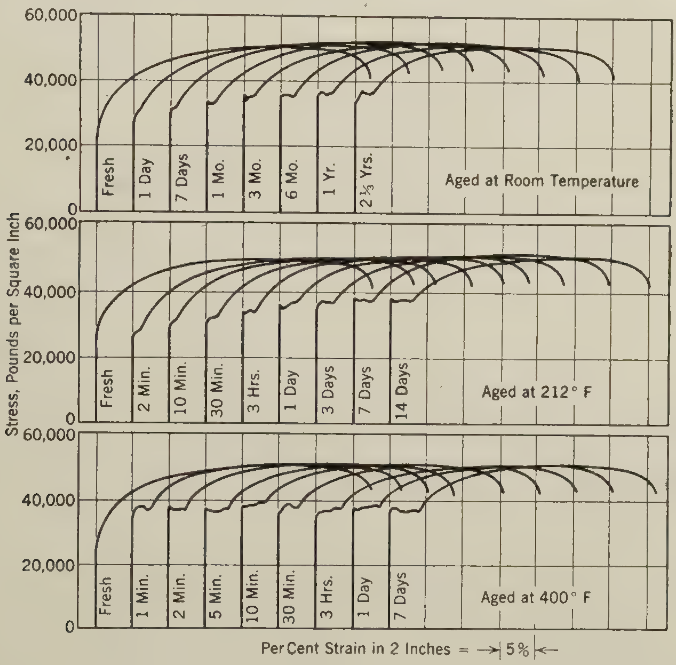

FIG. 5.—Stress-strain curves for mild steel sheets cold rolled 1% and aged for the times and at the temperatures shown. The steels that show a smooth curve will not stretcher-strain, those that show a jogged curve will. The jog at the yield point returns more rapidly at increased aging temperature. (Griffis, Kenyon and Burns.[5])

Copper Steels and Their Precipitation-hardening Heat Treatment. Copper forms no carbides and enters solely into the ferrite. Its behavior in improving the atmospheric corrosion resistance of steel and its strengthening effect upon ferrite are discussed in Chap. 5, Vol. II. Copper has little effect upon the martensitic quenching and tempering of steel. The effect it does have is analogous to the effect of Ni in this respect. But, if present in sufficient amount, it can be made to strengthen steel through a precipitation-hardening heat treatment.

The solid solubility curve of copper in ferrite[7, 8] is of the same type as that for cementite in ferrite, but the solubility is much greater. The exact location of the curve and the effect of other alloying elements upon its position are not fully known, but in general any ferritic steel containing from 0.70 to around 1.50% or more Cu is amenable to precipitation hardening, as shown in Fig. 6.

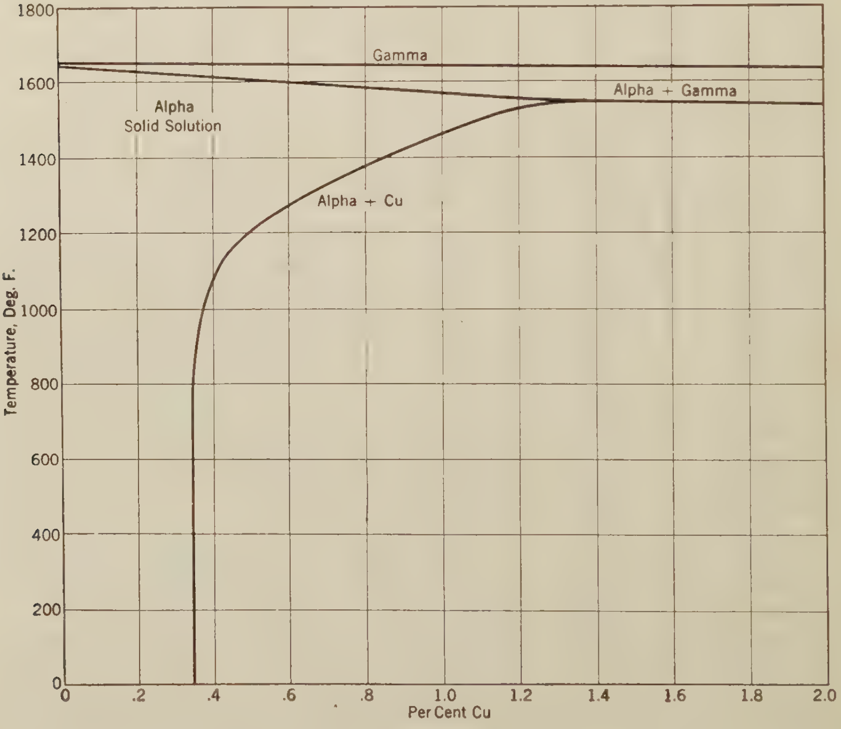

FIG. 6.—Solubility of copper in ferrite.

If such a steel is heated to 1475–1550°, i.e., above the solubility line, the Cu will go into solid solution; and, if the steel is then cooled, the Cu remains in solid solution. Upon reheating, say to 930° for 4 hr., the Cu (itself carrying what little Fe it can hold) is precipitated. The hardness and strength are increased. The changes in hardness with different reheating times and temperatures are indicated in Fig. 7. This phenomenon was first noted by Kinnear,[9] who applied it to cast steel. Low and medium C, Cu steels have their tensile and yield strengths increased by about 20,000 lb./in.^{2} by such treatment, the ductility decreasing at the same time, but less than when the increased strength is obtained in a normalized steel by increasing the C.

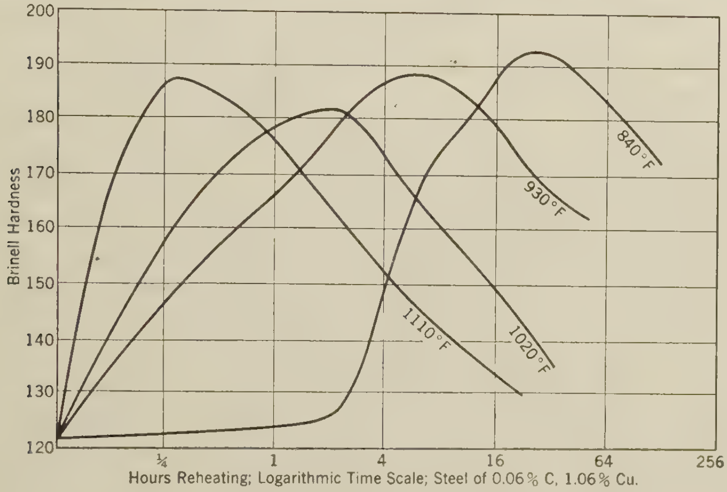

Fig. 7.—Hardness of a Cu steel, quenched from 1500°, reheated for the times and at the temperatures shown. (Smith and Palmer.[7])

Large Sections. The important feature from the point of view of heat treatment, is that the precipitation of Cu from solution is sluggish so that no severe quenching is required to hold it in solution ready for precipitation. A cooling rate as low as 1° per min. will retain it, so that even very large normalized sections are equally capable of throwing out the Cu at the center and at the edge upon reheating. Hence there is almost no mass effect with which to contend. Moreover, the precipitation goes on uniformly and gradually so that there is little tendency for distortion on reheating. Means by which the centers of heavy sections can have their strength boosted 20,000 lb./in.^{2} are rare, so that the potentialities of precipitation-hardened Cu steels are great.

It is not necessary to allow the material to get cold and then to reheat it. The steel can be normalized so as to produce fine pearlite from the Fe_3C, and when it has cooled a little below Ar_1 can be transferred to a furnace at, say, 930° and held there, when the precipitation of Cu will occur just as well as though the steel had been cooled and reheated to 930°.[10, 11]

Copper, like most alloying elements that enter ferrite, increases the strength in non-heat-treated steels, and it does not interfere with the usual heat treatment by quenching and tempering. This effect, in steels not intentionally precipitation hardened, is discussed in Chap. 5, Vol. II. Of course, the tempering, if below the solubility line of Fig. 6, involves some precipitation hardening; and the tempering time can be adjusted so as to oppose the sorbitic softening of tempering by some Cu precipitation hardening.

Effect of Composition. The precipitation-hardening effect is not marked in eutectoid steel, and is most effective in the lower C ranges. The effect begins to be evident at about 0.70% Cu, and is nearly as marked with 1% as with 2%. About 1.25% Cu is perhaps the optimum content when precipitation hardening is to be used.

The presence of other alloying elements, such as Ni, Cr, V, Mn, and Si, does not interfere with precipitation hardening by Cu. Hence a steel can be alloyed to produce a given set of properties on normalizing, and the precipitation-hardening effect of Cu superimposed upon those properties. Since this hardening involves some loss of toughness, it is advisable to reduce the C below that which would be used in its absence.

Properties Obtained. In order to bring out the improvement in properties due to precipitation hardening, data on steel castings in regular production, from Kinnear,[12] Zeuge,* and Finlayson † follow:

	Per Cent					Tensile	Yield	Elong. %	R.A. %	Brinell
	C	Mn	Si	Cu						
Kinnear..	0.29	......	......	0.94	N_1	88,000	58,000	19	31	179
					R_1	104,000	75,500	20.5	36.5	207
Zeuge....	0.26	0.73	0.34	0.97	Re	106,000	76,000	21.5	43.5	
Kinnear..	0.13	0.63	0.30	1.00	N_2	70,000	50,000	31	54	143
					R_2	89,000	70,000	23.5	49.5	196
Finlayson	0.11	1.04	1.23	1.74	A	81,000	64,000	31	56	
					ANR	106,500	88,500	25	53	

N_1 normalized from 1550°. R_1 normalized from 1550°, reheated to 950°.
N_2 normalized from 1600°. R_2 normalized from 1600° and reheated to 950°.
Re normalized from 1650°, reheated to 1000° F, held 4 hours.
A Annealed at 1740°—slow cooled. ANR—annealed at 1740°, slow cooled—reheated to, and air cooled from, 1580°, reheated at 915°.

* Sivyer Steel Casting Co.

† Private Communication, April 16, 1934.

Oxidation of Copper Steels. The presence of Cu sufficient for precipitation hardening involves one drawback. In the soaking pit or in heating prior to rolling or forging, the surface of steel containing, say, 1% Cu, is selectively oxidized, iron oxide forming more rapidly than copper oxide so that a film of Cu appears beneath the scale. If the steel is heated above the melting point of Cu the Cu tends to penetrate between the grains of the steel and produce a rough rolled or forged surface. Hence great care must be taken in controlling the heating temperatures. If a little Ni, say a third as much Ni as Cu, is present, this intergranular penetration is avoided, so high Cu wrought steels often contain some added Ni for this reason. Castings, which are not subjected to hot working, do not need the Ni addition.

The Ford Copper "Steels." The Cu addition is thought to lend fluidity to cast steel. The Ford Motor Company uses Cu[13] in so-called cast steel for crankshafts, brake drums, and pistons. These are cast from alloys of about 1.50 C, 1 Si, 0.80% Mn and from 1.50 to 3.00% Cu, with the highest Cu only in the brake drums. The brake drums also carry about 0.15% Mo, the crankshaft about 0.45% Cr, and the pistons about 0.10% Cr. These alloys are cast white, like malleable iron, and partially malleablized to give temper C and a pearlitic matrix. The castings are then normalized or annealed. The details for the crankshaft are: C 1.35–1.60, Cu 1.50–2.00, Si 0.85–1.10, Mn 0.60–0.80, Cr 0.40–0.50—P 0.10 max., S 0.06% max. Heat to 1650°, hold 20 min., air cool to 1200°, reheat to 1400°, hold 1 hr., furnace cool in 1 hr. to 1000°. Properties obtained are 107,500 tensile, 92,000 yield, 2.75% Elong., 2.25% R.A., 270 Brinell. The presence of Cu in these alloys is primarily to hasten graphitization in the short anneal for malleablizing and to add fluidity in casting.[14]

These high-C alloys are not precipitation hardened, since it would add little strength to the matrix, which is practically of eutectoid composition and hence too high in C to respond much. Work by Russel[15] has, however, indicated possible advantage in increased ability to withstand overstress in fatigue by carrying out the precipitation-hardening operation even though the static properties and the endurance limit are not improved.

Other Precipitation-hardening Steels. In nitriding steels (see Chap. 13 and Chap. 15, Vol. II) one may desire to back up the hard nitrided surface by a strong core. It would be convenient if the steel could be machined in soft condition before nitriding and hardened by precipitation hardening during nitriding. By utilizing the fact that Ni-Al steels may be precipitation hardened, this can be accomplished. Most of the nitriding steels contain Al anyhow. To such, the addition of Ni, as French and Homerberg[16] have shown, allows precipitation hardening to occur during nitriding.

A steel of 0.20 C, 0.50 Mn, 0.25 Si, 1.0 Al, 1.1 Cr, 0.25 Mo, and 3.50% Ni, after nitriding 48 hr. at 975°, has a yield strength of 115,000, whereas without the Ni and the precipitation-hardening treatment the yield strength of the core would be around 85,000. The hardness of the nitrided case is somewhat decreased by the presence of Ni, but its toughness appears somewhat improved. Ward[23] reports even higher properties for a steel of this class in nitrided parts for aircraft motors.

That the precipitation hardening is due to some compound of Ni and Al has been shown by Foley,[17] who produced it in a Ni-Cr steel to which Al was added. He found that elimination of the Cr did not affect the precipitation hardening, but leaving out either Ni or Al eliminated it.

Foley made a steel of 0.03 C, 0.16 Mn, trace of Si, 5.03 Ni, and 2.59% Al. By quenching from 1300° or above, the Ni-Al compound is held in solution. On reheating, as shown by Fig. 8, the Brinell hardness could be doubled. The reheating curves are of the type characteristic of precipitation hardening; 800° does not produce full hardness in a very long time, 1000° requires 24 hr., 1050° makes it fully hard in 4 hr. without decrease in hardness on longer holding, while 1100° also makes it fully hard in 4 hr., but further heating produces softening. Thus the 48 hr., 975° treatment used by French and Homerberg is entirely consistent with Foley's results.

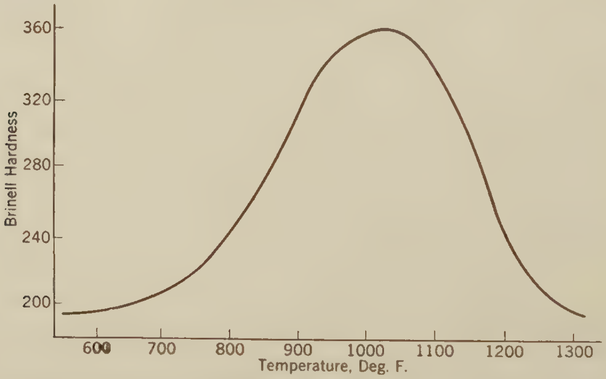

FIG. 8.—Change in hardness on reheating for 12 hr., at temperatures shown, a steel of 0.03 C, 2.60 Al, 5.00% Ni, after water quenching from 1300°. (Foley.[17])

That the same optimum precipitation hardening temperature of 950–1050° holds for the Ni-Al combination in the presence of other elements is shown by Foley's experiments on a valve steel of 0.50 C, 0.36 Mn, 0.94 Si, 1.58 Ni, 10.66 Cr, and 1.77% Al. The tensile strength of this steel was raised from 130,000 in the quenched state to 180,000 by reheating at 1000°.

Secondary Hardening. So-called "secondary hardening" of quenched and tempered steel, by which, instead of becoming softer with increase of tempering temperature, the steel hardens or at least maintains its hardness, is in some cases ascribable to precipitation hardening.

The strength of a 0.30 C, 2.60 Ni, 0.75 Cr, 0.50% Mo steel, oil quenched from 1650°, falls as the tempering temperature rises, whereas if 0.30% V is present the tensile strength remains practically constant to 1100° and the yield strength rises. Abram [18] shows the following:

	No. V		With V	
	Tempered		Tempered	
	850°	1100°	850°	1100°
Tensile	197,500	170,000	196,000	199,000
Yield	176,000	154,500	184,500	190,000
Elong. %	14	18	14	16
R.A. %	49	59	52	55
Izod	22	42	20	23

Abram ascribes this to the solution of vanadium carbide in austenite on the quench, its retention in solution in the ferrite formed from austenite on quenching, and its precipitation on reheating.

Cobalt, Molybdenum, and Tungsten in Precipitation-hardening Alloys. Various compounds that can be taken into and thrown out of solution will produce precipitation hardening. Among these are compounds of Fe with Mo or W. Carbonless alloys of Fe with 6 to nearly 40% Mo are precipitation hardenable. One of 30% Mo quenched from 2600° and reheated 20 hr. at 1155° rose from 40 Rockwell C to 65. The compound Fe_3Mo_2, or a phase (epsilon) of about that composition, is precipitated, according to Sykes.[19]

Sykes [20] has also shown that carbonless alloys of 8 to 33% W are similarly precipitation hardening through the separation, on reheating, of Fe_3W_2 or the epsilon phase of that series. A 25% W alloy quenched from 2730° and aged as shown in Fig. 9 gave a Brinell hardness of 185 to 370.

Complex steels in which large amounts of Mo or W appear, therefore, have possibilities of precipitation hardening by quenching to produce solution and reheating to produce precipitation of molybdides or tungstides. Owing to the hardness and stability of such compounds, alloys containing them offer promise as tools and dies for hot working. Cobalt, already used as a bond and matrix for sintered carbide tools, appears to serve as a solvent for such compounds.

Sykes [21] recently discussed the precipitation-hardening relations of Fe-Co-W alloys. Upon the basis of those phenomena and related ones in Fe-Co-Mo alloys, still more complex precipitation-hardenable alloys are being built up.

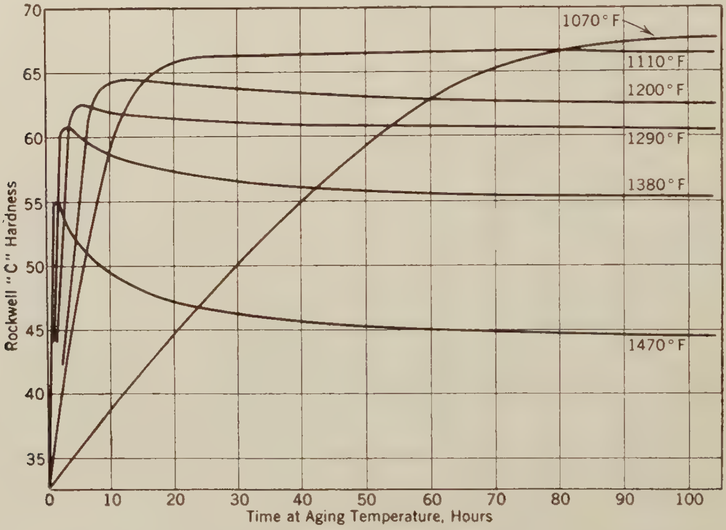

Fig. 9.—Typical age hardening or quench hardening behavior. Hardening of an alloy of 75 Fe, 25% W, after quenching from 2730° and reheating as shown. (Sykes.[20])

Harrington [22] describes an alloy of 1 C, 0.4 V, 6 Cr, 8 Mo, and 36% Co, remainder Fe, which, oil quenched from 1740°, shows 42 Rockwell C. Such an alloy is expected to be suitable for cutting tools of a general class between high-speed steel and sintered-carbide tools. With the Cr, V and C contents this alloy probably owes some of its hardness to carbides as well as molybdides.

Green * secures a similar range of hardness in an alloy of 0.12 C, 2 V, 3 Mo, 19 W, 30 Co, 0.20 Si, 0.30% Mn, remainder Fe, which

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* A. W. F. Green, Ludlum Steel Co. Private communication on grade 1548 Premet.

obviously owes its hardening power primarily to the tungstide rather than to carbide. In its treatment the forged stock is heated 5 hr. at 1625° and slow cooled, heated to 2250° for a very short time, quenched in lead at 1100°, then air cooled. This serves as the solution treatment and produces a machinable material of about 46 Rockwell C. Reheating to 800° for an hour hardens to 56 Rockwell C, to 975° for 20 min. to 66 Rockwell C, and to 1100°, for a still shorter time, to 71 Rockwell C. The exact treatment for best life of tools has still to be decided.

Permanent magnet alloys of the Al-Ni-Co groups, in which precipitation hardening is applied, are phenomenal in magnetic properties. (See Chap. 24, Vol. II.)

These examples will serve to indicate that whereas quench hardening through the formation of martensite can be applied to a much wider range of steels, the possibilities of precipitation hardening in those steels to which it applies are also of decided interest. The heat treater today needs to understand the principles of precipitation hardening as well as the principles of quench hardening.

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CHAPTER 9

GRAIN SIZE

Probably the greatest advance in theory and practice of heat treatment since the appearance of the previous edition of this book relates to the grain size produced when the austenite of a given steel is heated to different temperatures.

In earlier days it was recognized that some steels, especially high-C tool steels, varied in hardenability, even though their chemical analyses did not vary. The ability to harden readily was denoted by the term "body," or "timbre."

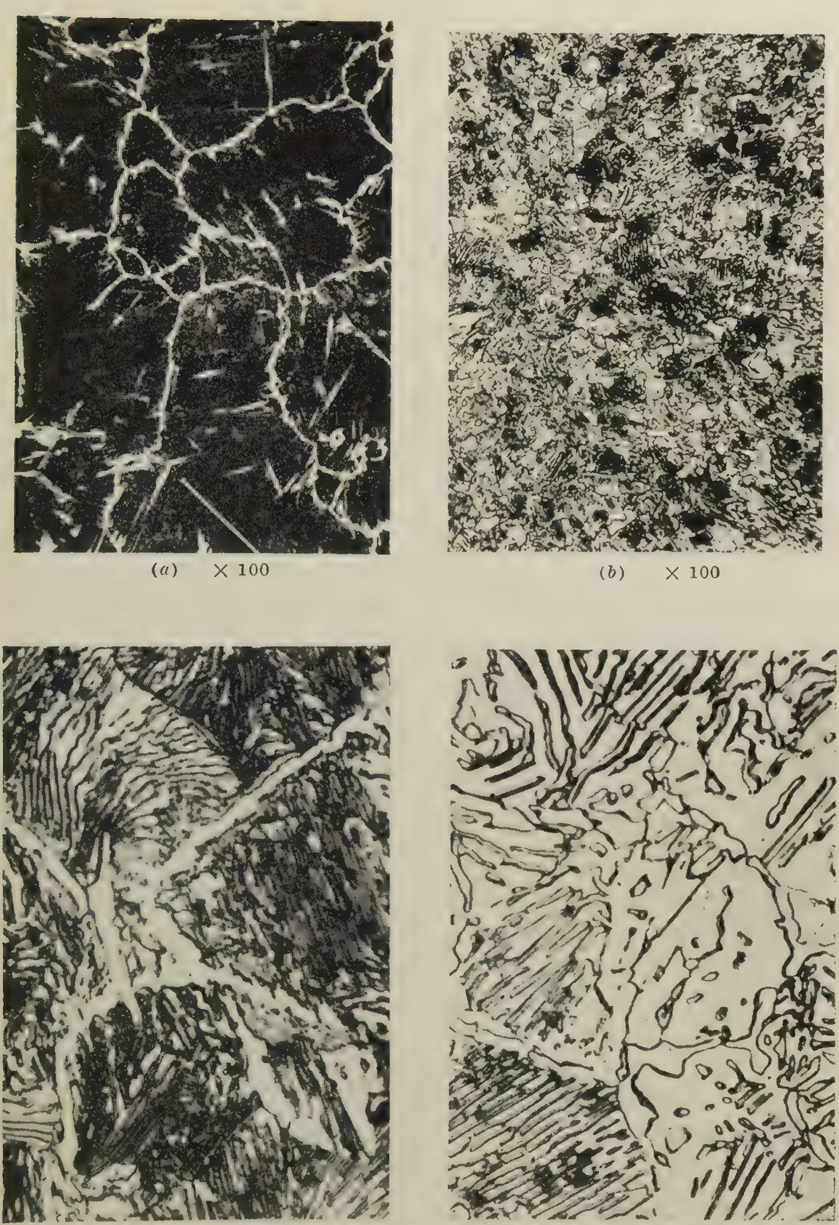
This variation in hardenability was known for many types of steels, but experimental attention was focused on it in relation to carburized steels. In quenching the high-C surfaces of box carburized steels in oil, or in warm water, it was found that some steels were prone to show soft spots, of primary troostite, instead of martensite. This means that the critical cooling rate for full hardening was greater than was locally reached. The cooling rate attainable was just on the ragged edge of that required, and a film of scale or a steam bubble adhering to the steel would throw the cooling rate of the steel beneath it, below that for martensitic hardening. More drastic quenching was required on such steels to avoid soft spots.

Normal and Abnormal Steel. The issue was somewhat confused by another phenomenon. The steels that gave soft spots had the cementite of the hypereutectoid zone coalesced into patches instead of outlining the grains as a network in the normal fashion, when the steel was cooled slowly from the carburizing temperature. The type with the network was called "normal," the other type, "abnormal" steel. At that time, because of the prevalence of soft spots in carburized "abnormal" steel, the implication that "abnormality" connoted inferiority was not disturbing and the nomenclature seemed natural. Further experience showed how to avoid the troubles with the "abnormal" steels and showed, moreover, that they had many virtues, so that today, while each type is superior in certain fields, the "abnormal" steels would probably be classed as so useful a product that some name implying superiority rather than inferiority would be chosen, were they to be re-named. Moreover the method of carburizing may affect the "normality," so that one should use the terms to refer to a structure rather than to a steel. This is brought out by Fig. 1.

The early studies were chiefly on carburized steels. In these, on sufficiently slow cooling from the carburizing temperature, the hypereutectoid cementite was found to outline what had been, at the carburizing temperature, very large austenite grains. Such steels were invariably deep hardening. Steels with fine grain (and, as later found out, whether their cementite tended to coalesce or not) were found to be shallow hardening. To evaluate hardenability, and to differentiate between normal and abnormal steels, the McQuaid-Ehn test was introduced. In time the idea of the effect of austenitic grain size became untangled from that of the distribution of the cementite, i.e., from the "normality," and grain size of austenite began to be studied as a major variable.

Experience with Vanadium Steels. At this early date grain size just happened, it was not intentionally controlled in ordinary C steels. However, there was a large background of experience with V steels, in which quite small amounts of V, 0.05 to 0.15%, were known to produce very tough, fine-grained, steels, this being especially noticeable in normalized steels where, in the absence of V, a high normalizing temperature was conducive to grain growth. Other alloying elements, particularly Mo and Ni, were known to have some inhibiting action on the grain growth of austenite.

Aluminum Additions. In those days final deoxidation of steel with a trace of Al, say an addition of 0.01 to 0.05%, was often employed to quiet a "wild" heat of steel. Need for Al was considered a sign of sloppy steel making and the steel makers who did use it were very loath to admit such practice. (Compare McQuaid.[1]) The melters would even provide Al at their own expense and use it to get sound steel against the instructions of the management. When Al was used, the steel had a decided tendency to be fine grained and also abnormal. Epstein and Rawdon[2] clarified the subject considerably. They subjected a series of steels, varying in "normality" by the McQuaid-Ehn test, to spectroscopic analysis for Al (which would not differentiate between Al present as the metal and as oxide) and found that fine grain and abnormality were connected with the Al content. An Al-free steel was coarse grained, normal, and deep hardening, whereas the Al-treated steels were the reverse. Epstein then added V and Al to a couple of ingots in a steel mill and compared these ingots with untreated ingots from the same heat. He found that Al acted very much like V in producing fine grain, and pointed out that steels having much the same properties as V steels could be made at



(*a*) × 100

(*b*) × 100

(*a*) Gas, × 1000

(*b*) Box, × 1000
(Nital etch.) (Lauderdale, Battelle.)

(*a*) Normal network structure. Gas carburized 8 hr., 1700°, slow cooled.

(*b*) Abnormal structure, carbide coalesced, Box carburized 8 hr., 1700°, slow cooled.

FIG. 1.—Normal and abnormal carburized structures in the same steel, after use of different carburizing media.

a lower cost, with controlled additions of Al. Commercial practice now utilizes Al very widely. A difference between the action of the two elements is that the V-treated steels are not prone to show the "abnormal" coalescence of cementite when carburized.

Later other strong deoxidizers such as Ti and Zr were appreciated as likewise capable of producing fine-grained austenite.

It has been suggested[44] that combinations of grain-refining elements used as final additions, instead of a single element, may allow improved control of coarsening properties, and facilitate the production of intermediate grain sizes at will. Little information is as yet available on the details of such practice.

The McQuaid-Ehn Test. We have spoken as though the austenite of the fine-grained steels was necessarily fine grained at any temperature in the austenite field and such an idea is more or less implied in much of the published discussion; but this is quite incorrect. The idea has sprung from the widespread use of the McQuaid-Ehn test * at one temperature, 1700°, only. Since most carburizing is carried out at 1700° or a little below, the test is adequate to show how the steels being examined will act in carburizing. When using the McQuaid-Ehn test to evaluate the behavior of a carburizing steel, the carburizing temperature, time, and medium that are to be used in practice ought to be adhered to. Houdremont and Schrader,[3] among others, have shown that on carburizing for the same time and temperature in a solid carburizer ($BaCO_3$ and charcoal) and in coal gas, several members of a series of steels studied behaved quite differently in regard to grain growth. The nature of the carburizing medium, therefore, cannot be disregarded.

Inadequacy of the McQuaid-Ehn Test. When the problem is not one of carburizing but of evaluating the steel as to coarsening propensities for other uses, the McQuaid-Ehn test cannot be considered as necessarily adequate, for; as Fig. 2 shows,[4] failure to coarsen at 1700° by no means ensures that the steel will not coarsen at a higher temperature, such, for example, as would be used in patenting. Nor does it tell whether a steel that does coarsen in the McQuaid-Ehn test might not have failed to coarsen at some lower temperature such as would properly be used in heating for quenching. Notwithstanding this obvious limitation and the fact that it would be more logical to apply the test at the exact temperature at which the steel is to be heated, if that is known, or over a range of temperatures if the object is to select the proper temperature, the "inherent austenitic grain

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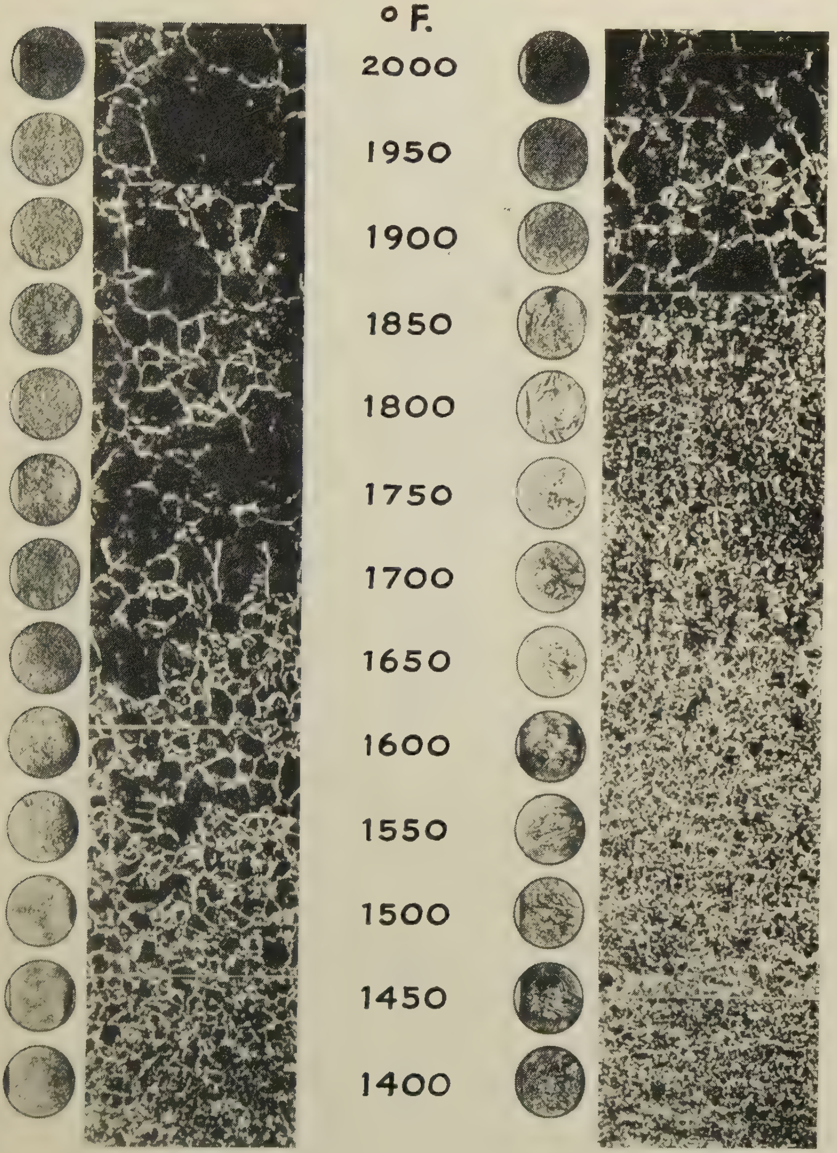
* Box carburizing with solid compound of the type that will produce a hyper-eutectoid case, for 8 hr. (usually at 1700°), followed by slow cooling.



(*a*) Coarse.
0.40 C, 0.71 Mn, 0.18 Si, 0.018 P, 0.029% S.

(*b*) Fine.
0.41 C, 0.76 Mn, 0.21 Si, 0.021 P, 0.035% S.

Fig. 2.—Structures of coarse- and fine-grained steels, normalized from a range of temperatures, and corresponding fractures when water-quenched from the same range of temperatures. The coarse-grained steel coarsens over the range 1550–1700°, the fine-grained one coarsens abruptly at 1900°. The quenched fractures are shown on a larger scale in Fig. 3, the steels are compared as to depth-hardening in Fig. 4, and Fig. 5 compares the properties of the normalized steels. (Schane.[4])

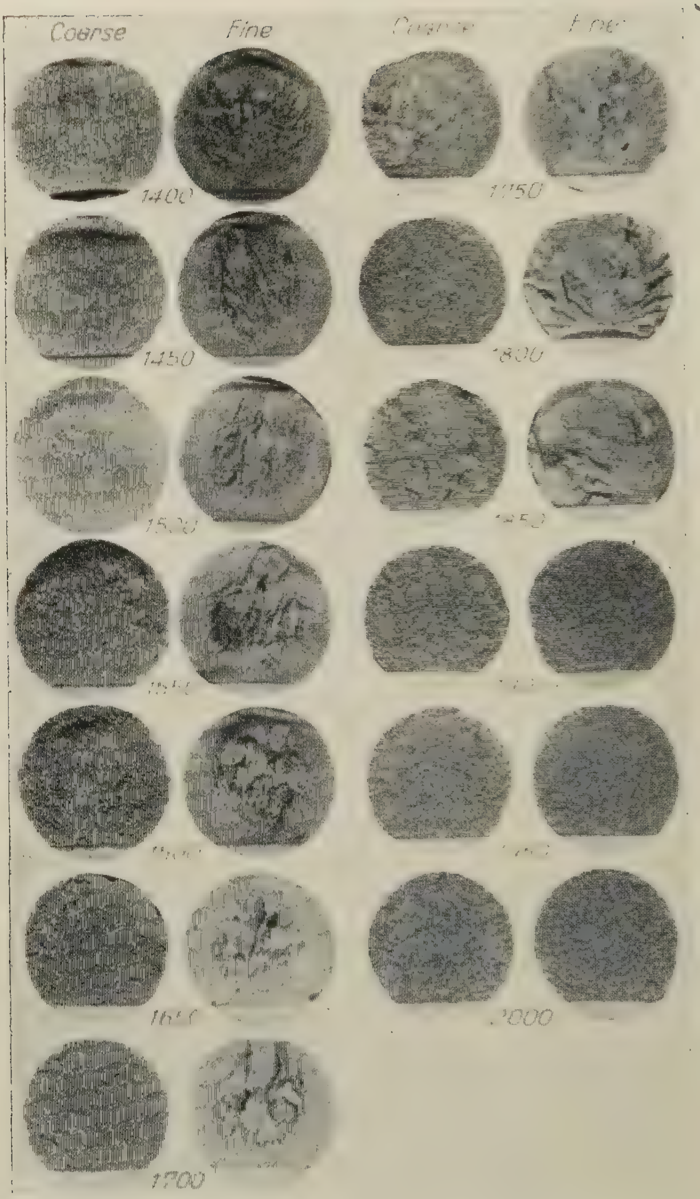


FIG. 3.—A larger view of the quenched fractures shown in Fig. 2. (Schane.[4])

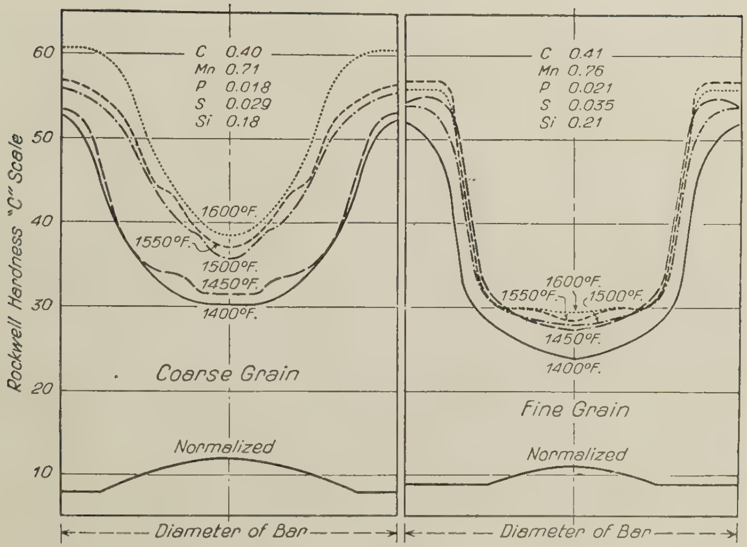


Fig. 4.—Hardenability curves on 1 in. rounds of the coarse and fine steels whose structures are shown in Fig. 2, and whose fractures are shown in Fig. 3. (Schane.[4])

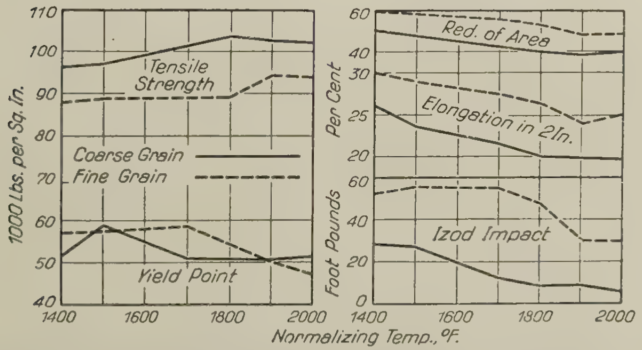


Fig. 5.—Mechanical properties of coarse and fine steels, normalized from different temperatures. The compositions are given and structures shown in Fig. 2. (Schane.[4])

size" is often allegedly described in terms of the grain size produced at 1700° only.

Standard Grain Size. The American Society for Testing Materials has set up a series of 8 grain sizes as seen under the microscope by which the steels are designated.* Figure 6 shows the size of these grains at × 100. In less precise terms, sizes 1 to 5 are classed as coarse and sizes 6 to 8 as fine. The grain size of the A.S.T.M. series varies in a regular progression, the number of grains per square inch on a section examined at × 100 doubling at each grain number. As Herty, McBride, and Hollenback[5] point out, it is the grain surface in three dimensions that is the real factor, and this varies as shown below.

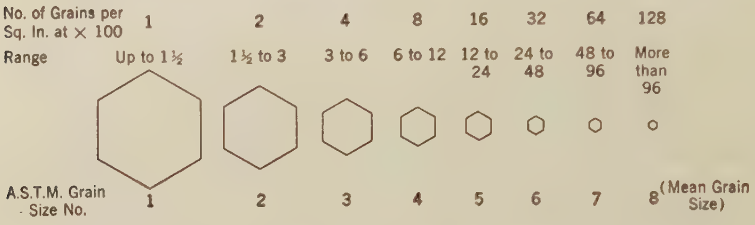


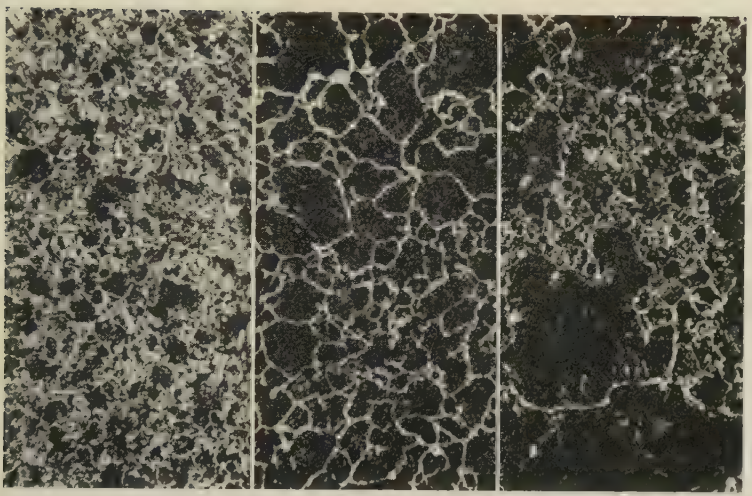
FIG. 6.—Idealized grain size of steel as viewed at ×100.

A.S.T.M. Grain No.	No. grains/sq. in. on section viewed at × 100	Grain[a] surface	Inter-[b] facial area
1	1	4.0	270
2	2	5.6	340
3	4	8.0	480
4	8	11.3	679
5	16	16.0	961
6	32	22.6	1360
7	64	32.0	1920
8	128	45.3	2720

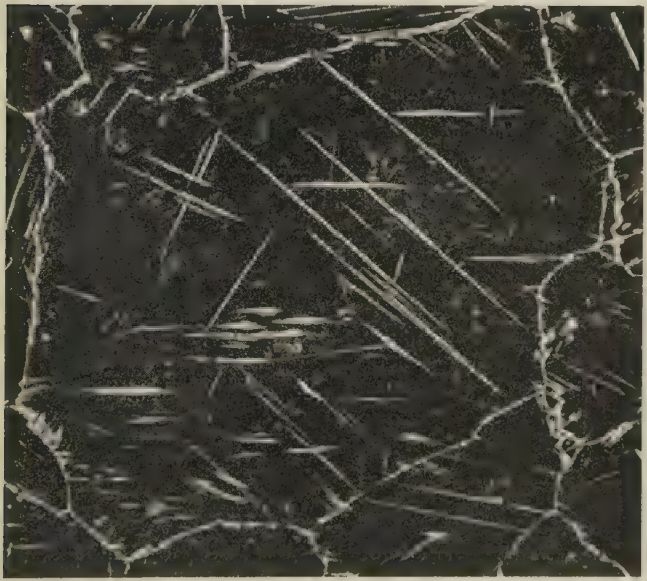
[a] Square inch of grain surface per cubic inch at × 100, calculated for cubical grains. Herty, McBride and Hollenback.[5]

[b] Mean total interfacial area square inch per cubic inch calculated for a 14-sided solid of maximum ability for close-packing. From Rutherford, Aborn, and Bain.[6]

* The official grain size chart of the American Society for Testing Materials showing hypereutectoid and hypoeutectoid zones of the case of steel subjected to the McQuaid-Ehn box carburizing test with solid carburizer and slow cooled to develop the cementite and ferrite networks, respectively, is given in A.S.T.M. Standard E-19-33. The chart is suitable for framing. This specification may be purchased from the American Society for Testing Materials, 260 S. Broad Street, Philadelphia, Pa., for 25 cents. A similar series of micrographs is shown in the American Society for Metals Handbook, the Handbook of the Society of Automotive Engineers, and Metal Progress, October, 1937, p. 435.



Fine Medium Duplex

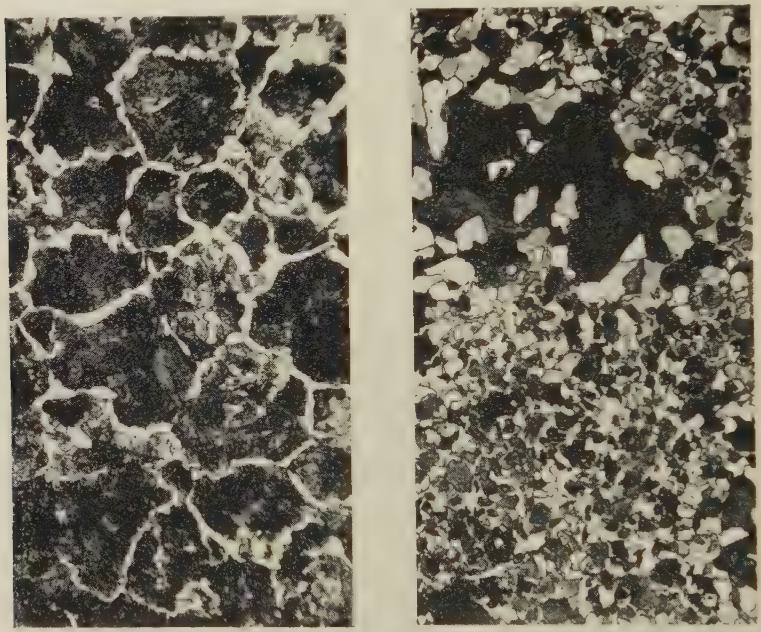


Very coarse. All × 100
(Lauderdale, Battelle.)

Fig. 7.—Hypereutectoid zones of different steels, all gas carburized 8 hr. at 1600° and slowly cooled. The grains are outlined by cementite.

The series may be extended at each end. Sizes of zero and —1 on the coarse end and 9 and 10 on the fine end are sometimes reported.

The austenitic grain size is fixed by the heating. It may or may not be clearly observable in the steel as prepared for service; but, if it is not, some method that will show that grain size must be adopted. The rate of cooling will alter the way the austenite grains are outlined, by cementite, ferrite or pearlite, according to the composition of the steel. Regardless of the way the steel is to be cooled for its regular heat treatment for service, the grain size established in its heating is fixed; various legitimate ways may be chosen to make this size apparent. In the McQuaid-Ehn test the composition of the steel is altered by introduced C, and the time required for C diffusion may be far longer than the heating time for any treatment except carburizing.

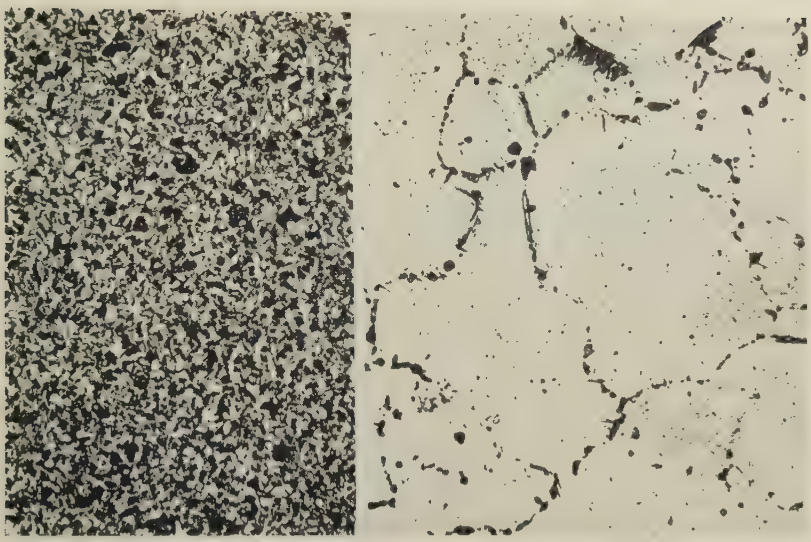

× 100 (Lauderdale, Battelle.)

0.47% C.
Coarse, 1800°.

0.35% C.
Fine and duplex, 1700°.

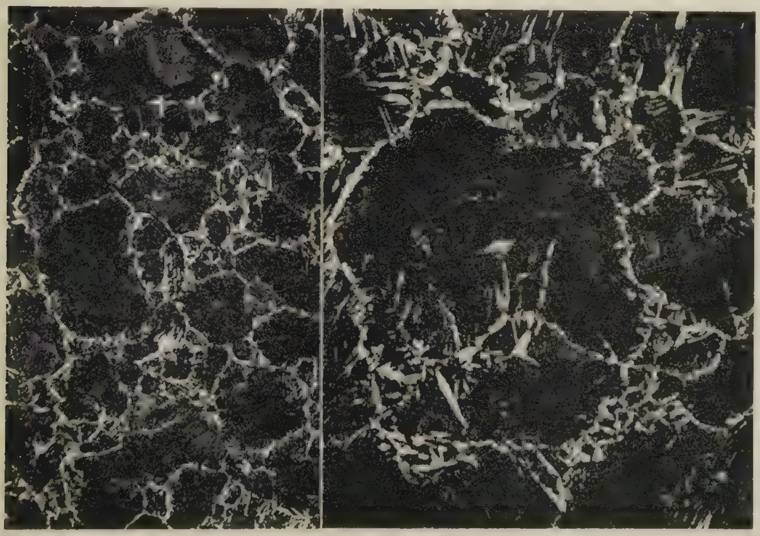
Fig. 8.—Cores of specimens box carburized 8 hr. at temperatures shown, and slowly cooled. The grains are outlined by ferrite.

Other Methods for Delineation of Grain Size. Slow cooling from the correct heating time and temperature will allow cementite of hypereutectoid (see Fig. 7) and ferrite (see Fig. 8), or pearlite of



Fine grain.
0.45% C steel. Heated 1 hr. at 1650°. Differential quench.

Coarse grain.
0.45% C steel. Heated 8 hr. at 1920°. Arrested quench at 1380°.



Medium grain.
0.41% C steel. Heated 1 hr. at 1650°. Differential quench.

Coarse grain.
0.41% C steel. Heated 1 hr. at 1750°. Differential quench.
All × 100 (Lauderdale, Battelle.)

FIG. 9.—Grain sizes delineated by quenching methods.

hypoeutectoid, steels to separate and outline the grains; but often it is better, with a hypoeutectoid steel, to cool down into, but not past, the critical range and then quench. This is termed an "arrested quench." (See Fig. 9.) At compositions near the eutectoid some other means must be found. Some conditions of quenching and tempering can usually be found that will serve.[7] If one end of the specimen is water quenched, the other left out of the water, the zone between the hardened and unhardened parts will have pearlite nodules outlining the grains. This is termed a "differential quench." (See Fig. 9.) Light tempering of fully hardened steel to secondary troostite may also be used. In any case, a suitable etching reagent must be found that will show contrast between grains and boundaries.

Quenching from the temperature in which one is interested, tempering lightly, and etching to develop the martensitic grain size that reflects the austenitic grain size, is described by Bain and Vilella,[8] and is much to be preferred to a test at an arbitrary temperature.

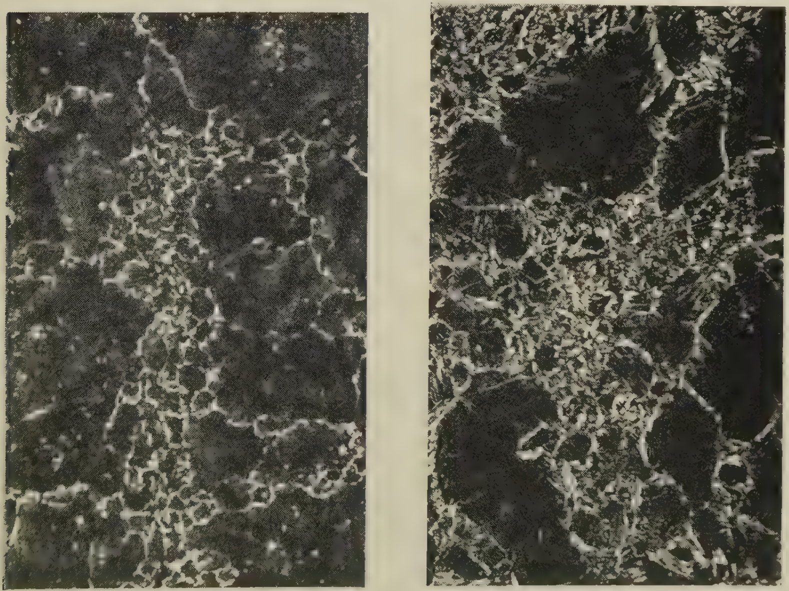
Heat tinting is sometimes employed to bring out the grain boundaries. Eutectoid steels and some alloy steels may not readily etch or heat tint in such fashion as to show the austenitic grain size clearly, so that it is occasionally necessary to work out a combination method. Tobin and Kenyon[9] carry out the heating under slightly oxidizing conditions so as to attack the grain boundary material and render it more strongly attacked in the etching process.

In the case of high-speed tool steel, which is usually very fine grained, Snyder and Graf[45] suggest examination at × 1000 and counting of the grains intercepted by a diameter of the field, after etching. Such steels, in a condition for service, may have the austenitic grain size revealed by etching with up to 10% HCl and 3% HNO_3 in methyl alcohol. This method allows discrimination among steels whose grain sizes correspond to from 8 to 10 on the extended A.S.T.M. scale, which if compared visually at × 100 with the usual chart, could only be classed as "finer than grain 8." The cutting properties are influenced by variations in this range of grain size.

Since, in hypoeutectoid steels, ferrite, and in hypereutectoid, cementite, is rejected at the grain boundaries of austenite as it passes through the critical range, a pure grain size effect is only present in eutectoid steels. In others the lining of the grain boundaries with a different material has its effect. Herty and co-workers[10, 5] point out that the rate of cooling through the critical affects the amount of boundary ferrite or cementite. With hypoeutectoid steels the amount of ferrite correlates closely with the impact resistance. So, although it is ordinarily phrased that coarse-grained and fine-grained steels behave thus and so, this means only that the cited behavior holds for the steels cooled at the certain rate being discussed.

Three-dimensional Grain Size. Since the grain count or estimate is made on a plane section viewed at × 100 it is obvious that this plane will cut the mass of crystals so that some crystals will show their maximum cross section while others will show only a small cross section where the plane cuts across a corner of the crystal. Hence "mixed grain," as Rutherford, Aborn, and Bain [6] show, is not evidenced unless the smaller grains are grouped in clusters. When the cross section merely shows smaller grains rather regularly interspersed among the larger grains, the effect is due to the geometry of the section. When true mixed grain appears it is an indication either of inhomogeneity of the steel or that the coarsening process was caught in an active state, coarsening having begun but not proceeded to equilibrium for the temperature used, so that the absorption of small crystals by larger ones was irregular. Figure 10 shows mixed grain.

Coarsening Characteristics. It is thoroughly illogical to try to evaluate all the grain size and grain growth characteristics of a steel by a test under one condition only, as by applying the McQuaid-Ehn test at 1700° alone. A figure that tells more is the temperature range



1.10% C tool steel, held 8 hr. at 1700°. Slow cooled, nital etch.

0.41% C steel, held 1 hr. at 1650°. Differential quench. × 100 (Lauderdale, Battelle.)

Fig. 10.—Duplex, or mixed, grain size.

of coarsening, but that figure likewise fails to tell the whole story. More complete data, including the austenitic grain size at each of a range of temperatures below and above the temperature at which notable coarsening begins, are necessary to describe the "coarsening characteristics." The situation is comparable to endurance testing. One cannot make any broad comparison of steels on the basis of life at one load; a series of tests at different loads is necessary to establish the S-N curve.

The term "inherent austenitic grain size" is a misnomer, for no single grain size is inherent. It may occur that two steels treated the same way and having, under that treatment, the same grain size will behave differently and in ways that seem to be related to the fact that one of the steels would have coarsened and the other would not, had they been heated to a higher temperature. That is, the nature of the steels, which is evidenced by variations in ability to coarsen, also produces other variations in behavior even though the coarsening tendency has not come into play.

Depth Hardening *vs.* Grain Size. The effect of grain size that we are most often interested in is upon the critical cooling rate, or in other words the depth of hardening. Instead of trying to approximate this from the single-temperature McQuaid-Ehn test, or even carrying out that test at a range of temperatures, Shepherd [11] and the Swedish Jernkontoret quench at a range of temperatures, and fracture the specimens. The grain size is evident on visual inspection and can be quantitatively estimated by comparison with known standards. The depth or penetration of hardening is observed on sectioned and etched samples, or from the fracture. According to Shepherd [11] it is feasible to evaluate the hardening behavior of steels on a "P-F" (penetration-fracture) basis as he has described in detail.

Determination of Hardenability. Hardenability of alloy steel is so vital a matter that two important recent papers discuss an approach toward standardization of methods for its determination. Jominy and Boegehold [13] discuss a test for evaluating the hardenability of the case on a carburized steel. They spray water at 75° onto the bottom of a test specimen 1 in. diam., 2¾ in. high, held in a suitable fixture so that the spray strikes the bottom face only, the sides being protected by a shield that prevents any water striking them. Or, the piece may be machined so that the bottom has a small projecting ring to serve the same purpose. The specimen may be carburized 8 hr. in compound at 1700° and tested direct or oil quenched, reheated to the desired quenching temperature (with protection against scaling of the bottom face as by resting it on a graphite disk), quickly placed in the quenching fixture, and quenched. The surface hardness along the side is then determined by Vickers diamond or Rockwell C. The specimen is then ground to a depth of 0.015 in. and another set of hardness readings made. The distance above the quenched face at which 600 Vickers Brinell or 60 Rockwell C is obtained indicates the depth hardening propensity of the steel. (Since the outside of the carburized surface is usually softer than the layer beneath, the more deeply penetrating Rockwell impression gives a higher reading than would be expected for the Rockwell C—Vickers relation for homogeneous steels.)

The cooling rate of the surface of the cylinder when the bottom face is quenched can be found by spot welding thermocouples to it, and with that information the critical cooling rate of either high-C carburized or ordinary C and alloy steels can be determined.

Striking differences between coarse and fine-grained steels were shown. Steel of 1.25% Mn hardened to 600 Brinell for 5⁄16 in. up the specimen when it had a grain size of 4 to 5, and only for 1⁄16 in. when it had a grain size of 7 to 8. A steel of 0.70 Mn, 1 Ni, 0.70% Cr (grain size 6) hardened 2½ in. The greater hardenability of steel quenched from a high temperature was shown by comparison of steels tested direct from 1700° or oil quenched, reheated to 1425°, and tested.

Variations in hardenability of a dozen lots of 1.75 Ni, 0.25 Mo, 0.40–1.60% Mn steel, all fine grained (6 to 8) were found, the vertical distance of full hardening ranging from ½ in. to 2½ in. Some of this variation can be accounted for on the basis of grain-size variation within the 6–8 range, some to the Mn content, but not all, so it is evident that there is a need for a more direct evaluation of hardenability than is given by chemical analysis and the determination of grain size.

Since the cooling rate at the center of various sections is known from the work of French,[14] the Jominy-Boegehold test can be converted into a determination of critical cooling rate required to harden a given steel, when the test is made on an uncarburized sample, since the distance hardened in the test can be converted into the section in which a given steel will harden throughout. Obviously decarburization in heating for quenching would have to be prevented to make the test accurate for such a purpose.

Another Method. Burns, Moore, and Archer[15] harden a 4-in. length of 1-in. round bar in water and cut it in two (half way up) with an abrasive wheel. The 1-in. diam. face is then Rockwelled along 8 radii on circles 1⁄16 in. apart. The readings are plotted (or recorded and averaged in a way that is equivalent to plotting) so as to give (1) surface hardness, (2) area under the curve in the plot, (3) center hardness. The area is reported in terms of "Rockwell inches." Thus the result can be expressed in three numerical values e.g., 62–46–30, meaning that the surface hardness was 62 Rockwell C, the area 46 Rockwell inches, the center hardness 30 Rockwell C. Although such a set of figures does not tell as much as the actual curve, it does offer a means of describing hardenability in words, over the telephone for example, when the curve cannot be transmitted. The two extremes of the curve are stated but the shape of the curve is not implied in the numerical rating, since curves of various shapes could give the same area. However, a coarse-grained steel will show deep hardening and a large area, while a fine-grained one will show shallow hardening and a smaller area. The quenching temperature and the quenching medium would have to be stated in a broader application of the test.

From the surface hardness of the 1-in. round, the area, and the center hardness, i.e., the "S-A-C" rating, these authors expect it to be possible to estimate approximately the depth hardening curve for larger sections, though the relationship of the curves would have to be established for steels of a wide range of hardenability before this would be very accurate. Figure 11 shows curves for three carbon steels, all fine grained. The efforts, described above, at getting quantitative hardenability figures indicate that more precise evaluation of this property than is given by grain size alone is needed.

Grossman [46] emphasizes the value of actual hardness testing across a section of the hardened piece, as contrasted with the mere visual examination of the fracture, and points out that differences in hardenability are more evident when the examination is made on sections near the critical size. In an example cited of two steels, differing but slightly in hardenability, both gave hardness curves very nearly alike in 1-in. diam., where both hardened almost to the center, and very nearly alike in 1½ in. diam., where the hardness gradients were steep on each, but at 1⅛ in. diam., the center hardnesses and the gradients differed materially.

Methods for Control of Grain Size. The steel maker [16, 17] is much interested in the methods of control of grain size and the mechanism of the inhibition of growth. Epstein, Nead, and Washburn [18] have discussed the former, and the latter has been very widely discussed, without final agreement on just what the mechanism is. McQuaid [19] lays emphasis on the Al present as metal in Al-treated steels rather than on the Al oxide, supposed to be present in submicroscopic form. Brophy and Parker [20] think that both Al oxide and excess

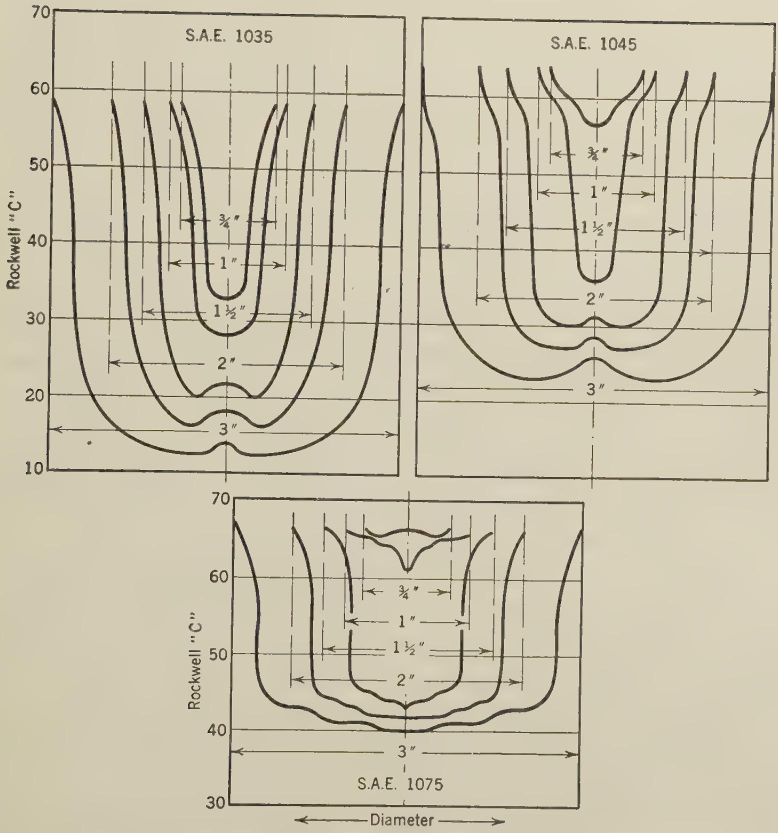


FIG. 11.—Effect of diameter on depth of hardening in water-quenched, fine-grained steels. (Burns, Moore and Archer.[15])

Composition, %

	S.A.E. 1035	S.A.E. 1045	S.A.E. 1075
C	0.37	0.48	0.78
Mn	0.78	0.78	0.77
Si	0.23	0.27	0.19
P	0.016	0.016	0.028
S	0.032	0.030	0.031

of metallic Al affect the behavior of steel in heat treatment. Some open-hearth superintendents also believe that both are required in producing steels of controlled grain-coarsening characteristics. Foulkes[21] expresses this point of view. Derge, Kommel, and Mehl[22] studied high purity Fe containing Al, Si, Al_2O_3, and SiO_2, and concluded that as high as 0.20% Al or Si in solid solution did not inhibit grain growth, whereas Al_2O_3 or SiO_2 did inhibit it. They state: "These results are all in agreement with the hypothesis that fine-grained austenite is the result of inhibition of grain growth by properly dispersed insoluble particles." Most authorities agree with Derge, Kommel, and Mehl in attributing the inhibition of grain growth primarily to mechanical action. In the case of V it is assumed that V-bearing cementite is very difficultly soluble and that, until the austenite has dissolved it and attained homogeneity in C content, grain growth does not occur. Solution and homogenization must first be accomplished. If they have been accomplished, for which a temperature well above those ordinarily used for hardening or for normalizing is required, grain growth is rapid. It takes intentional overheating to coarsen such a steel. In the case of alumina (and perhaps aluminum nitride) or titania and zirconia (and their nitrides) it is seldom thought that these non-metallics have appreciable solubility, though their solubility may turn out to have an important bearing on the problem.[12]

Sulfur as a Possible Factor. Another factor that may enter, as suggested by Sims,[23] is the S content. There is strong evidence that, in cast steels, the type of inclusion formed depends on the relative amounts of Al and S, and that some of the Al is combined in the form of an aluminum sulphide. Hence, with a given Al addition, the Al should divide itself among Al oxide, Al sulphide, and uncombined metallic Al, according to the amounts of O and S present, and the temperature. The Al sulphide itself may be one of the non-metallics that obstruct grain growth, and it would be expected to have a higher solid solubility than Al_2O_3. Titanium and Zr also form sulphides as well as oxides, carbides, and nitrides. It is quite common to add both Al and Ti for grain-growth control. Since the effective non-metallic compounds are submicroscopic, it is practically impossible to prove just what compounds are present, to say nothing of their exact quantity. For the present, at least, the results can only be described in terms of the amount of addition made to steel at some particular stage of deoxidization that corresponds to a more or less definite content of reactive O, S, N, etc., instead of in quantitative terms of percentage of a certain growth-obstructing constituent.

Probable Mechanism of Inhibiting Action. The action is considered as a mechanical inhibitor to growth which may appear and disappear according to its solubility relations. The known effect of thoria in W lends color to this view.

Schane[24] points out that the inhibiting substances have some degree of solubility in austenite and the solubility and rate of diffusion vary with temperature. At lower temperatures the substances diffuse to, or are rejected to, the grain boundaries and prevent growth. At some higher temperatures the boundary film is dissolved or partly dissolved and ruptured so that it no longer prevents grain growth. Dorn and Harder[12] have expanded the idea of solubility of inhibiting materials.

If any of these various kinds of submicroscopic, non-metallic particles are present in the austenite grain boundaries they would act as mechanical barriers to the growing together of adjacent austenite grains, much as an unfluxed oxide film will interfere with welding. At some sufficiently elevated temperature the adjacent grains should be active enough to overstep this barrier, engulf the inhibiting particles, and proceed with grain growth. Actually, at some very elevated temperature this sudden growth does occur in the fine-grained steels, in contrast with the more gradual growth of the grains in a steel free from Al, V, or other inhibiting elements. This is evident in Fig. 2.

It might be pointed out that the submicroscopic alumina inclusions postulated are not necessarily accompanied by the clumps of alumina often found in poorly made Al-treated steels. The fine-grained steels are not necessarily any "dirtier" than the coarse-grained ones.

Hardenability as Influenced by Grain Size. With this brief and incomplete picture of the possible mechanism by which steels may show differences in austenitic grain size, we may turn to a discussion of the results in heat treatment that flow from those differences. Whereas steel of intermediate grain sizes can be procured on the market, control to produce the intermediate sizes is not yet sufficiently advanced and sufficiently widespread to put these into very wide commercial use. Coarse-grained material, say 1 to 3, or fine-grained, say 6 to 8, can be readily secured and our general discussion will relate to these two classes rather than to one specific size.

Whatever the mechanism of the inhibition of grain growth of austenite in the fine-grained steels, and whether or not grain size is the cause of the effect, or merely a concomitant effect of a more deep-seated cause, the practical result is that, other things being equal, when a steel is cooled from an austenite temperature at which the grain is coarse, the steel is deep hardening; when it is fine, it is shallow hardening (though fine-grained, deep-hardening steels can of course be produced by adding a suitable alloy to a steel base that would be fine grained and shallow hardening).

Sluggishness of Coarse-grained Austenite. In terms of time required to transform all austenite to pearlite at 1250°, on cooling, when a 1.17 C, 0.30% Mn steel is quenched into a fused bath at that temperature, Davenport and Bain[25] find that all the austenite is gone in 6 sec. when the steel was initially heated to 1550° and had a grain size of 8, i.e., was very fine. Coarsening it by heating to 1750°, to a grain size varying from 2 to 6, caused the austenite to take 30 sec. to transform completely, i.e., 5 times as long.

Thus we see that coarse-grained austenite is sluggish, and much easier to cool down past Ar′. This makes the coarse-grained steels deeper hardening than the fine. Figure 4 shows Rockwell hardness across cylinders of 1040 steels of the same composition, (outside of their tiny content of Al), but varying in grain size, quenched as shown. The same thing is shown by the depth of the hardened zones on the fractures in Fig. 3.

Effects on Quenching Temperature. Grain size is thus an index of hardenability, and many of the statements in Chap. 6 as to critical cooling rates in relation to C content, etc., have to be modified. They are strictly true only when steels are compared that are of equal grain size at the temperature chosen from which to cool. Thus the effect of the maximum temperature of heating before quenching upon the hardenability can be related to the grain size produced by that heating. The repeated warning not to heat far above the end of Ac_3 for quenching, a sound one for steels that coarsen rapidly just above that temperature, is not necessary and often not advisable for steels that remain fine grained at much higher temperatures. Thus the fine-grained steels have a much wider safe hardening range, their heat treatment is less fussy and more fool proof.

Effects on Normalizing. An obvious application for the use of a grain-refining addition, like Al or V, is in steels to be normalized whose critical cooling rate is such that they tend to air harden. Cases in point are Mn steels with C and Mn both on the high side.

Cast steel that is to be normalized instead of annealed may show higher strength and lower ductility than is normally desired. For example, steel of 0.30-0.35 C, 1.50-175% Mn, normalized, may show 110,000 tensile with but 15% elong. and 20 ft.-lb. impact. Suitable addition of V, Al, or Ti so as to make the steel fine grained will drop the tensile to 95,000, raise the elongation to more than 25%, and approximately double the impact. The grain-coarsening effect of Mn and the tendency toward hardening at a low critical cooling rate characteristic of coarse-grained steel are corrected by refining the grain.

An analogous application is in wrought Mn-bearing steels to be welded. Cooling from the welding heat, in thin sections, tends to produce some air hardening in, and adjacent to, the weld, with consequent brittleness. Several of the low-alloy, high yield-strength steels (see Chap. 11, Vol II) utilize 1 to 1.40% Mn as chief alloying element and, in order to secure the desired strength, the C may go up to 0.25-0.30%.

Such steels are not considered suitable for welding unless they can be heat treated after welding. However, the limiting C and Mn contents for safe welding without subsequent heat treatment are not definitely known.[26] The variations in reported behavior may readily be due to unrecognized differences in grain-coarsening propensities. It is fairly certain that Mn-V steels of this class can be safely welded at a C + Mn content that would be unsafe without the addition of 0.05-0.10% V. Other grain-refining additions would probably have an analogous effect.

Effects of Grain Size on Mechanical Properties. Fine-grained hypoeutectoid steels, in cooling through the critical at rates that produce pearlite, do not coalesce the ferrite into a connected network surrounding the pearlite grains as readily as do the coarse-grained steels. A thick, connected, ferrite network leads to a low impact resistance, its absence to a high one. This, as well as the finer grain in itself, makes the fine-grained normalized steels show high impact resistance, a fact long appreciated in respect to normalized V steels, whose outstanding characteristic is their fine grain.

Data from Schane,[4] assembled in Fig. 12, show that in whatever condition, annealed, normalized, or quenched and tempered, S.A.E. 1040 is superior in ductility and vastly superior in impact when it is fine grained.

Rolf [27] also gives data for 0.40% C steels, of coarse and of fine grain, water quenched from 1525° and tempered at 800° to 1200° by 100° intervals. The tensile and yield strengths of the fine grained steel were 5 to 10% less than for the coarse grained, but the ductility and toughness were increased. At the lowest draw temperature the elongation and reduction of area were increased 20 to 25% and the impact resistance was tripled. Similar, but less marked, superiority in ductility and impact resistance was found at the other tempering temperatures.

Scott [28] finds, for very hard steels, spring temper and harder, that a similar effect holds, as is shown in Fig. 13.

It is universally conceded that, for service at ordinary temperatures, a more desirable combination of strength and toughness is had in fine-grained than in coarse-grained steels. This is most clearly shown by the impact test, and since the serviceability of the fine-grained steels is universally recognized from actual performance, increased understanding of the effects of grain size has brought an increased appreciation of the value of the impact test.

Bain [29] shows that, whether quenched and tempered, or austempered, a coarse structure leads to poor impact (un-notched bars) as

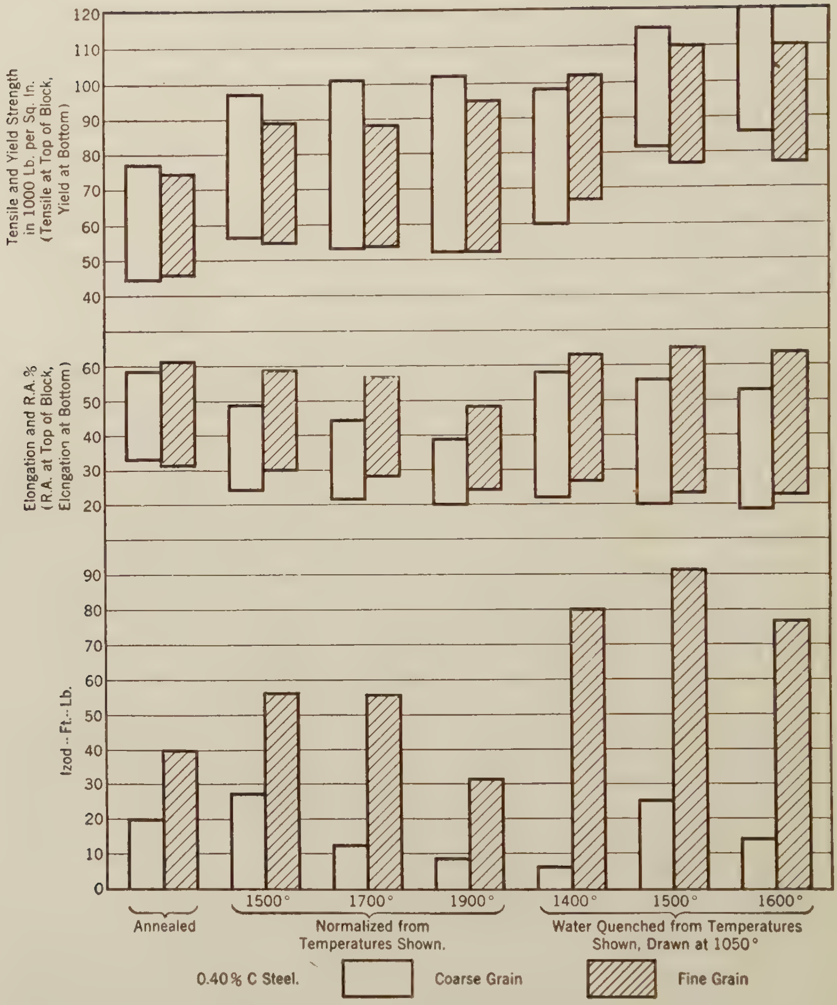


FIG. 12.—Effect of grain size on mechanical properties. (After Schane.[4])

indicated in Fig. 14. Since the hardness of all specimens was the same, the effects shown are due to grain size and structure.

Grain Size *vs.* Machinability. Low resistance to impact seems to be connected with easy machinability in rough machining. Hence a coarsening of austenite by heating above the coarsening temperature before normalizing is often used to produce a more readily machinable steel; after rough machining the steel may be reheated to a lower temperature at which the austenite does not coarsen, to produce a finally tough steel on the next normalizing or on quenching and tempering.

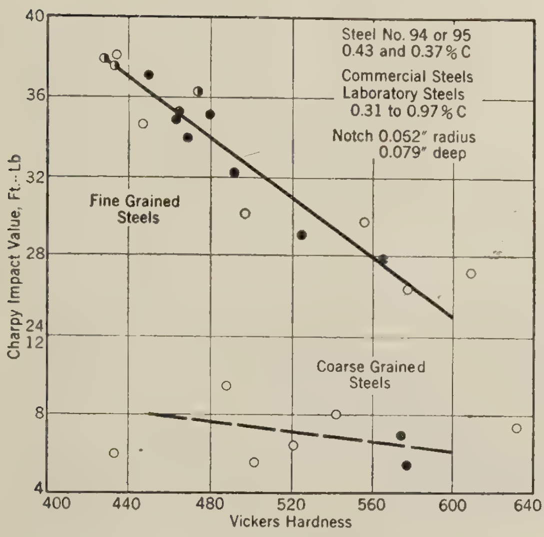


FIG. 13.—Impact values of C steels quenched from 1525°, tempered at 300–570°, to hardness shown. (Scott.[28])

A fine-grained steel gives the smoothest surface on fine machining, so, if we are seeking finish instead of ease of machining, the fine-grained steels are superior.

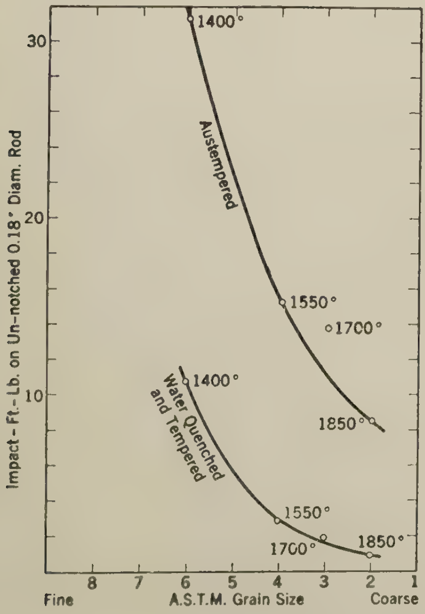


FIG. 14.—Grain size vs. impact (un-notched). The same steel, containing 0.85 C, 0.42 Mn, 0.18 Si, 0.022 S and 0.008% P, was used in all cases, but was given varying austenitic grain sizes by heating to the different temperatures shown. All specimens were cooled to 1400° before quenching. One series was quenched in molten lead and not drawn (i.e., austempered). The other series was water quenched and then drawn. All specimens had the same hardness, 50 Rockwell C. (Bain.[29])

Distortion and Cracking. As further drawbacks to the coarse-grained steels, and added reasons for not heating to the coarsening temperature before quenching, are the greater distortion on quenching and the greater propensity toward quenching cracks.

With these differences in behavior in heat treating and in properties after heat treating it is obvious that the heat treater can attain his ends not only by selection of a steel of suitable chemical composition but also, and almost equally, by selection of one with the desired grain size characteristics.

Juggling the Grain Size. By choosing a steel whose austenitic coarsening temperature is midway in the feasible range of temperatures for industrial use, he can make it coarse grained or fine grained at will, by heating to a temperature above or below the coarsening temperature. The possibility is also opened up of using controlled grain size C steels instead of alloy steels, or of attaining a desired result with a simple alloy steel of controlled grain rather than with a complex or more highly alloyed one.

With the production of steels of controlled grain-growth tendencies and with the understanding of the phenomena of grain growth, these variations can be utilized with certainty of results. This is a great advance over the situation we were in when the grain size of a heat of steel was an accident and when the heat treater could not be sure that a steel of the same chemical composition as one he had previously used would respond similarly to treatment.

Availability of Controlled-grain Steels. Instead of having to order a steel by its brand name, and hoping that the steel-maker, by adhering to the same routine in melting and finishing the heat, will consistently produce that brand with the same grain-coarsening and hardenability characteristics, the heat treater can now order his steels with these characteristics specified. The large users, such as the automotive manufacturers, have done this for some time back. For example, Peterson [30] mentions the grain size as prominently as the S.A.E. number in his description of steels for automobile gears, springs, axle shafts, and the like. The grain-coarsening properties are considered as important as the chemical composition.

The small user is in a more difficult position, since he ordinarily orders in small lots, often from a warehouse or jobber rather than directly from the steel maker. He may, therefore, have to deal with a bewildering succession of steels which he supposes will react the same in heat treatment because they bear the same S.A.E. number, but which actually vary in response to heat treatment, not only because of variation within the S.A.E. range of composition but also even more so from variation in coarsening characteristics. One large warehouse firm now takes a whole heat from a steel mill, identifies the bars by stamping, and determines the exact chemical analysis, grain size by McQuaid-Ehn test, inclusion rating, and response to heat treatment, showing tensile properties and hardness for a selected quenching temperature and draw temperatures. Thus the user may adjust his treatment (compare p. 199, Chap. 7) to the requirements of the particular lot of steel and may order subsequent lots with more certainty of duplication of their behavior.

Hardenability of Tool Steels in Respect to Grain Size. Tool steels have long been considered to show such mysterious variations in behavior of different lots that purchase by brand name was the rule. However, the mysterious "body" of C tool steels has been stripped of most of its mystery by the application of knowledge of grain size effects. For example, Digges and Jordan [31] studied two 1% C steels, essentially identical in chemical analysis for all elements usually determined, i.e., differing chemically only in that one contained 0.01% Al instead of a scarcely detectable trace, but which showed entirely different depth-hardening properties. They found the differences thoroughly explained by the grain size, grain-coarsening characteristics, and resultant differences in critical cooling rates.

Schempp and Shapiro,[32] working with 1.05% C steels, found that cold working, and to some degree, prior heat treatment, alter the grain-coarsening characteristics. In the discussion of this paper it was clearly brought out that other factors, such as those mentioned by Schempp and Shapiro, may be superimposed on the "inherent" tendencies of the melt itself, to modify the behavior of the finished steel. Portevin [33] has discussed the effect of cold work.

Palmer [34] brings out the fact that in high-C tool steels, depth of hardening and propensity toward grain growth do not necessarily go hand in hand. Moreover, the fracture of a hardened specimen quenched from a normal temperature, say 1450°, may not reveal differences in two tool steels (they may both appear fine grained and show equal depth of hardening), so he advocates quenching in brine from 1550° to reveal the differences. This increase in quenching temperature is more likely to bring out the innate differences. The one that shows coarsening from overheating in such a test will be inferior in toughness even when quenched and drawn at the usual temperatures, the variation in toughness often being of the order of 100% when evaluated by the torsion impact test, while the hardness of the two steels is identical. Differences are also apparent in susceptibility to grinding checks. The more readily coarsened and more brittle steel will have tiny surface cracks when abused in grinding; the tougher one will stand more abuse.

Coarsening Propensities More Vital than Actual Grain Size. Boulger* has also found instances in medium-C steels where the impact resistance of lower-C steels falls off at a temperature approaching the coarsening temperature but still too low to show more than a very

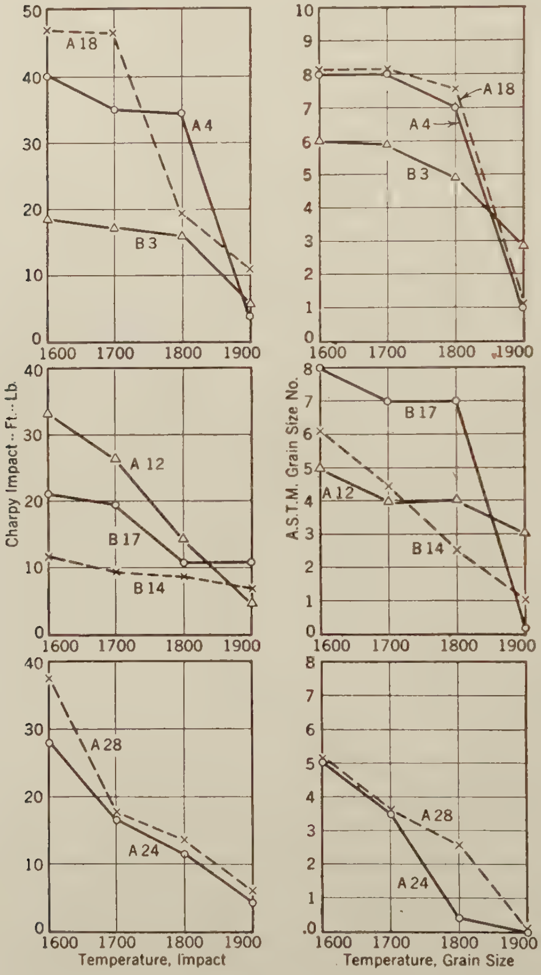


FIG. 15.—Comparison of impact and grain size of annealed steels.

Compositions, %

No.	C	Si	Mn
A_{18}	0.18	..	1.00
A_4	0.20	..	0.45
B_3	0.45	..	0.55
A_{12}	0.17	0.06	0.51
B_{17}	0.42	.	0.84
B_{14}	0.50	0.26	0.80
A_{28}	0.08	..	0.39
A_{24}	0.19	0.20	0.91

All steels held 8 hr. at temperatures shown and slowly cooled. (Boulger, Battelle.)

little coarsening. Thus in Fig. 15, from his work, the shape of the impact and grain-size curves is similar, taking into account the initial differences in impact due to chemical composition; but in the samples of A12, A18, and B17 annealed at 1800° a drop in impact occurs that is disproportionately great in relation to the degree of coarsening at 1800°.

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* Unpublished work, Battelle Memorial Institute.

Grain Coarsening Propensities *vs.* Creep. Notwithstanding the close relation between grain size and impact resistance, steels of the same analysis and of the same grain size do not necessarily have the same impact resistance. Data reported by the Joint Committee[35] indicate that *at a given grain size* in two steels of similar composition, the one initially fine grained and requiring a higher coarsening temperature for grain growth may have a much higher impact resistance than the one with the lower coarsening temperature—as shown below.

Steel No.	C	Composition, %			S	Al Added %	Normalized from	Grain Size	Coarsening Temp.	Izod Impact	Norm. from	Grain Size	Izod
		Mn	Si	P									
54	0.49	0.76	0.20	0.017	0.028	0.025	1650°	2 to 4	1650°	4			
61	0.50	0.75	0.22	0.020	0.026	0.025	1650°	2 to 4	1650°	4			
52	0.46	0.54	0.23	0.016	0.030	0.105	1800°	2 to 4	1625–1775°	18	1550°	6 to 8	27
56	0.51	0.75	0.21	0.018	0.033	0.10	1850°	2 to 4	1700–1800°	16	1550°	6 to 8	23

No.	Norm. from	Tensile	Yield	Elong., %	R.A., %	Grain Size	Creep—Rate of Deformation, % per Hr., at End of 500 Hr. (Unless Noted) at 850° at 20,000 lb./in.^{2}
54	1650°	100,000	49,000	22.5	43.5	2–4	0.00023
61	1650°	107,500	53,500	20.5	40.5	2–4	0.00016
52	1800°	92,000	51,000	25.5	47	2–4	0.00048
52	1550°	91,500	56,000	27.5	51	6–8	0.00910 at 140 hr.
56	1850°	106,000	57,500	21.5	42	2–4	0.00019
56	1550°	102,000	59,500	25	49	6–8	0.00415 at 140 hr.

As the table shows, a difference of 2 or 3 to 1 in creep rate occurred in steels of the same grain size and similar composition, while the difference between the rates of fine- and coarse-grained classes was 20 to 1.

The steels 54 and 61, coarsened to a grain size of 2 to 4 from 1650°, showed only 4 Izod against about 25 Izod for fine-grained steels, but 52 and 56, coarsened to a grain size of 2 to 4 at 1800-1850°, gave around 17 Izod. That is, in the same coarse grain size, the steels of low coarsening temperature lost 80 to 85% of their resistance to impact whereas those that completed their coarsening at higher temperature lost only 30 to 35%.

This is of particular interest because steels with sufficiently coarse grain to show the better resistance to creep at high temperature (see Chap. 19, Vol. II) that accompanies coarse grain size are prone to have very low impact at ordinary temperatures, and a means by which both creep resistance and impact resistance may be maintained at desirable levels would be of great value.

The creep resistance in a pair of lower-C steels, in fine-grained condition, was decreased tenfold by doubling the Al addition, even though the coarsening temperature was not reached and both were normalized from the same temperature to the same grain size.

	%C	Mn	Si	S	P	Al Added %	Coarsening Temp.	Actual Normalizing Temp.	Grain Size	Creep, Rate of Deformation at 850° and 15,000 lb./in.^{2} % per hr.
2809	0.29	0.68	0.32	0.045	0.011	0.025	1825°	1650°	8	0.00059 at 500 hr.
2810	0.28	0.62	0.31	0.041	0.011	0.05	1875°	1650°	8	0.00580 at 190 hr.

	Tensile Properties—Grain Size 8 Norm. from 1650°				Charpy
	Tensile	Yield	Elong., %	R. A., %	
2809	77,000	50,000	32	61	34
2810	76,500	52,500	33	63	37

The steel in which the actual normalizing temperature more closely approached the coarsening temperature gave the better creep result.

Grain Size an Important Variable. A duplex structure might have the poor creep properties of its fine grain and the poor impact properties of its coarse grain. It would appear that creep properties in uniformly fine-grained steels may be better the nearer the fine-grained ones are brought to their coarsening temperature without coarsening (though the creep properties resulting from still higher temperature treatment may be better); and impact properties in uniformly coarse-grained steels may be better the less the coarsened ones are taken above their temperature for uniform coarsening.

Coarsening propensities, i.e., the temperatures of the beginning and end of coarsening and the range of temperature over which duplexing occurs, need consideration as well as mere grain size.

The differences in behavior of coarse- and fine-grained steels at high and low temperatures indicate that they may perhaps be more justly ascribed to "inherent" propensities of the steels which are *also* indicated by grain size and grain-coarsening temperature rather than being directly due to the grain size itself. The grain size may be a concomitant rather than a cause. At any rate before one can justly compare the effects of two alloying elements or two percentages of the same alloying element in steel, one should first make sure that the steels compared are comparable in respect to grain size and grain-coarsening propensities.

Practically every statement about the heat treatment of and the properties of heat-treated C and alloy steels needs to be modified by a statement of the grain size and the coarsening properties of the steel, or at least to be read while carefully holding in mind the modifying effect of variations in austenitic grain size.

Adjustment to Secure Advantages. By choosing an initially fine-grained steel whose coarsening temperature is not too high, one may make it either fine or coarse grained at will by adjusting the normalizing temperature. Each type has its advantages and disadvantages, well summed up by Wilson[36] as follows: "In general, the coarse grained steels are recommended for good machinability, intense hardening, easy coarsening of structure and high temperature service above 950°. The fine grained steels are recommended for economical case hardening, good toughness, superficial hardening of the direct hardening steels, small warpage in hardening, best mechanical properties of untreated steel and high temperature service below 950° F."

It is not correct to set 950° as the dividing line in high temperature service, as this dividing line undoubtedly varies with the composition of the steel, as well as its grain size. In plain C steels the dividing line is probably quite a bit lower, but the general principle of existence of a dividing line of temperature above which coarse grain has superior load-carrying ability seems well established.

The effect of grain size in steels for high and low temperature service will be discussed in Chap. 19, Vol. II, but some comment on low temperature behavior may be made here.

Grain Size *vs.* Low Temperature Impact. Since impact resistance increases as the grain size becomes finer and, as was brought out on p. 52, Chap. 1, impact resistance tends to become low at low temperatures, the effect of grain size at very low temperatures is of particular interest. While both the alloying elements and the grain size have an effect on low-temperature behavior, a paper by Kinzel, Crafts, and Egan[37] and its discussion show that the beneficial effect of fine grain holds down to very low temperatures, thus corroborating the results of Herty and McBride. Herzig and Parke[38] studied a series of alloy steels down to − 200°. While there were considerable differences in impact resistance of fine-grained steels of different compositions, those with coarse grain, or of duplex structure containing some coarse grains, were much poorer than those with fine grain. For low-temperature impact resistance, fine grain is as clearly an advantage as it is a drawback for high-temperature, load-carrying ability.

Wire and Forgings. Other features in which coarse- and fine-grained steels differ in behavior are constantly being noted. Boynton[39] prefers coarse-grained wire for cold drawing to high tensile strength, but fine-grained for patenting. Vignos[40] finds that coarse-grained steels make better finished forgings, and Ellis,[41] in a study of forgeability, concludes that the larger the grain size the more readily steels are deformed in forging.

Grain Size in Deep Drawing Steel. The importance of grain size in deep drawing steels is universally recognized. A clear description of the requirements is given by Winlock and Leiter[42] who point out that too coarse a grain size produces rough "orange-peel" surfaces and too fine a one predisposes to stretcher straining. The best balance between the two evils is, they state, found at A.S.T.M. grain size 7.

Commercial Importance. It is stated[43] that 75% of the killed steel now made, is made under grain size control.

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SECTION II. SURFACE-REACTION PROCESSES

CHAPTER 10

CASE CARBURIZING

Introduction. So far, we have been concerned with the effect of heating and cooling upon steel whose composition at the surface has been assumed to be the same as that of the body of the piece, that is, the surface has not been intentionally or unintentionally altered.

An important branch of heat treatment is concerned with the intentional alteration of the surface so as to give it hardness and wear resistance, by introducing C (carburizing) or N (nitriding) or both (cyaniding). Of equal interest is the introduction of precautions in heat treatment that will avoid undesired chemical changes of the surface at the elevated temperatures necessitated in heat treatment. These changes may result in the oxidation and scaling of the surface, or in the removal of C from it (decarburization).

All these effects involve chemical reactions at the surface of the steel, which are controlled by adjusting the composition of the steel, the composition of the medium in contact with it, and the temperature and time of contact.

The Purpose of Carburizing. High-C steels, of eutectoid and hypereutectoid composition, like rail steel and C tool steel, quenched and given a low temperature draw, are hard and wear resistant, but lacking in toughness. Because of their brittleness they would not stand up in some types of gears, for example.

Wear resistance is required only upon the surface, and for a short distance below, in machine parts, for when only a thin layer has been worn away, the part fails to fit and must be discarded anyway. The core of the piece does not need hardness as much as does the surface; it might far better be tougher and less hard, though the required core hardness varies with the service.

In order to produce the needed combination of a hard surface and a tough interior, the process of *case carburizing* is invoked, by which a tough, relatively low-C steel has its surface layer enriched in C. (Compare Fig. 4, Chap. 2.) The case-carburizing process, and a few other analogous surface-hardening methods like cyaniding and nitriding, are the only ones carried out by the heat treater in which he intentionally alters the chemical composition of the steel he uses, before he starts to govern its physical properties by heat treatment. Case carburizing, the manufacture of a high-C steel layer upon a low-C steel, precedes case hardening, the heat treatment of the case, or of both case and core, by quenching and tempering.

Carburizing Done in the Austenitic State. Case carburizing is a chemical process by which C in an active state is brought to the surface of the steel, absorbed by the steel, and allowed to diffuse inward to the desired depth. Active C is supplied by the decomposition of carbon monoxide (CO) from solid carburizers, or of some hydrocarbon, in gas carburizing. Before discussing the differences, we should first consider matters that are common to both methods. Ferrite has almost no power to dissolve C, whereas austenite has the ability to dissolve large amounts. For this reason carburizing is carried out above Ac_3, with the steel austenitic, not ferritic. Practically no carburization is performed on ferrite, though Kinzel[1] has suggested how this might be accomplished. Just as the low-C ferrite patches and the higher-C pearlite patches, transforming into austenite on heating, tend to become homogeneous through diffusion, so does a high-C surface layer upon austenite tend to diffuse inward. If more C (in an active form) is constantly supplied to the surface, this transfer can be kept up until finally the whole steel is of practically uniform, high C content. In very early days high-C tool steel was made, utilizing such a "cementation" process. Very deep carburizing is used in the production of one type of armor plate. We are usually concerned only with the production of a relatively shallow surface layer.

Types of Steels Used. Several considerations dictate the C content of the steel used. The rate of diffusion is greater the greater the difference in C content between outside and inside. For core toughness, also, the ductility of low-C steels is sought. Hence the bulk of the steels used runs from about 0.10 to 0.20% C. When machining properties prior to carburizing are important, the higher end of the range is used, since the lower C makes the steel gummy and more liable to tear.

However, if it is essential to give the case better support than the soft steels will give, higher C, around 0.35% may be used to allow the core to be strengthened and hardened by heat treatment. High-C tool steels are sometimes heated in carburizing media in order to produce a very thin skin of higher C content than the body of the metal, but more commonly in order to avoid decarburization. When tool steel is heated for hardening, packed in solid carburizers, the process is called "pack hardening," while the solid carburizer process on usual steels is more commonly called "box carburizing," or "pack carburizing."

Alloys may be used in the steels to hasten the rate of C pick-up at the surface, accomplished by the addition of carbide-forming elements, especially Cr or Mn, or to alter the response to heat treatment of case or core or both. Nearly all the alloying elements are usable for one purpose or the other, save Si in large amounts. Silicon is a graphitizer and decreases the rate of absorption of C, and the stability or the carbide. The steels used are listed in Chap. 11, and their selection will be discussed in Chap. 14, Vol. II.

Grain Size in Carburizing. Coarse-grained steels carburize slightly more rapidly than fine-grained ones. (Fig. 1.) As was brought out in Chap. 9, the coarse-grained steels are more readily hardenable, i.e., they require less drastic quenching to harden, while a fine-grained case may be just on the border line of hardenability so that adhering gas bubbles in quenching cause soft spots beneath them. The fine-grained, "abnormal" steels were once avoided for this reason. At present, fine-grained steels are ordinarily specified for carburizing, a suitable hardenability being secured by adding a little of one or more depth-hardening elements, Cr, Mo, Mn, or Ni.

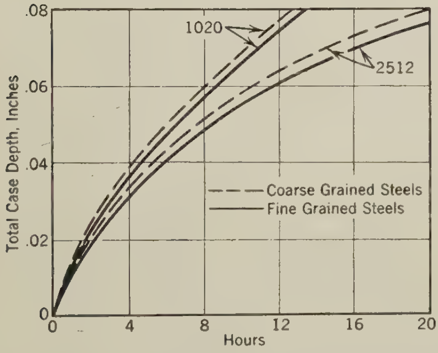


Fig. 1.—Case depth produced on plain C (1020) and Ni (2512) steels in solid carburizer in the times shown. (Int. Nickel Co.)

The main reason for sticking to fine-grained steels is their superior toughness, both in case and core. With a coarse-grained steel it is ordinarily necessary to employ four heating operations, as will be discussed in Chap. 11, one to carburize (during which grain coarsening occurs), followed by slow cooling; another above both Ac_3 and Acm to refine the grain of the core, and take excess carbide into solution, followed by quenching; another above Ac_1 to refine the high-C case, followed by another quench; and a final low temperature draw to relieve quenching stresses.

With fine-grained steels one quench usually suffices and, with suitable steels, this may be carried out from the carburizing temperature. The fewer the heatings and coolings, the less the warpage.

When rapid carburization is desired and a high carburizing temperature is employed, the use of a steel whose grain coarsens badly at the necessary temperature may involve so much embrittlement, and so much warpage in the complicated heat treatment necessary to repair the embrittlement, that for measures of economy as well as quality the fine-grained steels are much superior.

Treatment Prior to Carburizing. The only finishing operation possible after carburizing and hardening is grinding or lapping, the surfaces being too hard to machine; so the pieces are machined to size before carburizing. If much stock is to be removed in finish grinding a correspondingly thick case must be produced. Ordinary grinding allowances are of the order of 0.01 in. for heavily carburized stock and where warpage has not been avoided. On small pieces, with careful control of all conditions, the allowance for grinding may be as low as 0.002 in.

The preliminary machining of soft steel requires that it be of proper composition and grain size to machine quickly and cheaply, with the proper surface finish. The very low-C steels that might be chosen from the point of view of core toughness are likely to tear on machining, so the C content chosen is often a compromise.

The fine-grained steels that give the best core toughness rough machine less readily than coarse-grained steels. Hence there is a demand for steels that will be fine grained at the carburizing temperature but whose coarsening tendency is great enough to allow coarsening for machinability.

The stock is then normalized from a very high temperature so that the steel does coarsen and is suitable for rapid machining. Many fine-grained steels will coarsen at 1800°, but some require still higher temperatures.

In order that the stock will not warp from internal stress after machining, upon being heated to the carburizing temperature, as well as to ensure that its propensity toward grain growth is the same in all parts, stock not already in stress-free condition may have to be normalized or quenched and tempered prior to carburizing. Such treatments also make the stock have the same sort of machinability from piece to piece, whereas hot-worked bars or forgings finished at different temperatures, may not have.

Unless it is desired to coarsen the grain intentionally for machinability the treatment for uniformity will be a normalizing from the usual temperature for the particular steel being used, or a quench from the usual temperature followed by a high draw, say at 1100° for C steels, and somewhat higher or lower for some of the alloy carburizing steels.

McQuaid[2] points out that the carburizing steels with an air-hardening tendency, especially those of high Ni, accompanied by Mo or Mn, may be difficultly machinable if air cooled for machining prior to carburizing. In such cases rapid cooling to 1200°, with slow cooling to 500°, or a reheat to 1200° and slow cooling to draw some of the hardness conferred by air cooling, may be required. The ordinary C and low alloy carburizing steels are, however, suited to the usual normalizing practice.

Prevention of Carburization. Since the carburized case is unmachinable after hardening, any parts to be machined or drilled after carburizing must be kept soft and uncarburized, unless the work is cooled from the pot and the machining done before hardening. This is not possible in single-quench work. When the area to be left soft is large it is possible to pack a pot with inert material at the places to be left soft and active carburizer at those to be hardened, but it is difficult to keep the carburizing gas away from the places covered with inert material. Holes or other places where such coatings will remain in place can be packed with a fire-clay, water-glass cement.

Isolated spots to be kept soft are best protected by Cu plating them. The parts to be hardened are "stopped off" with an insulating coat, like baking japan, or a lacquer that will not strip during plating, the places to be coated are cleaned as meticulously as though it were a regular plating job and those places plated with Cu from a cyanide bath. Or the whole piece can be plated prior to finish machining and the Cu machined off the portions to be hardened. If the Cu plate is sufficiently thick, pore free, and adherent, it will protect the surface that it covers from carburization.

Temperature. In order to make the steel austenitic and to have the temperature high enough so that diffusion will go on fairly rapidly, yet not so high as to cause excessive grain growth, the carburizing operation is carried out around 1700°, the usual range being from 1650 to 1750°.

Influence of Time and Temperature on Depth of Case. The rates of C absorption and diffusion vary with temperature; these processes go on more rapidly at first and then slow down. The case depths resulting from different times, temperatures, and carburizing media follow curves like those of Figs. 1-7. Curves from Bramley [3] given by Williams [4] and shown in Fig. 8, as well as those of Giolitti,[5] indicate

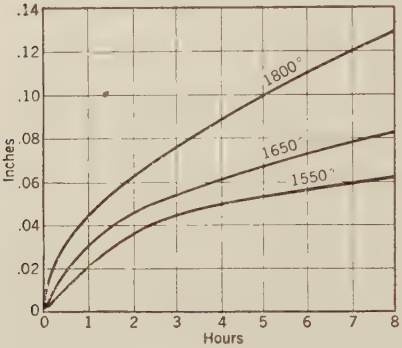


FIG. 2.—Rate of carburization in ordinary solid compound. Case depth vs. time and temperature.

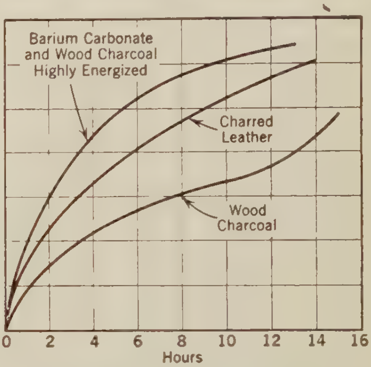


FIG. 3.—Relative rate of carburization with different solid carburizers at 1650°. (Scott.)

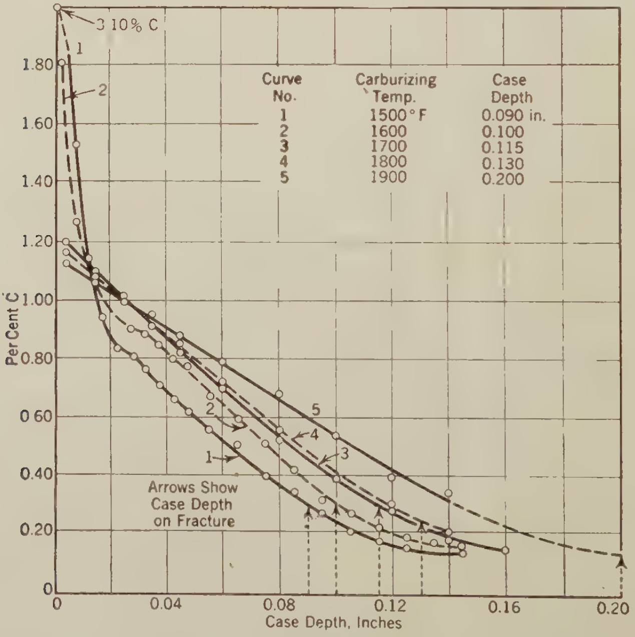


FIG. 4.—Case depth vs. C content, showing the effect of carburizing temperature on C concentration, in box carburizing at temperatures shown. (Schlumpf, quoted by McQuaid.[29])

the importance of diffusion. If carburizing is done at a low temperature, the exterior of the case is very high in C; it may even be so high as to exceed the solubility of C in austenite at that temperature and form undissolved carbide particles, but just below the surface the C is low. As the carburizing temperature is increased, diffusion is more rapid and C is pulled away from the surface so fast that it does not build up there as high as at lower temperatures. Alloying elements affect the rate of absorption of C at the surface and the rate of dif-

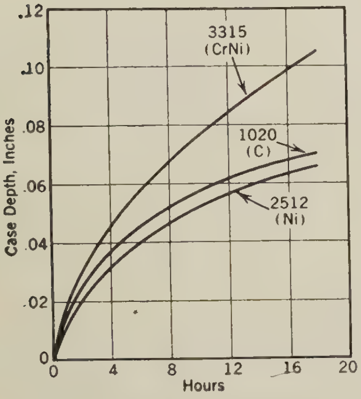


FIG. 5.—Rate of gas carburizing of three steels at 1700° (A. S. M. Handbook.)

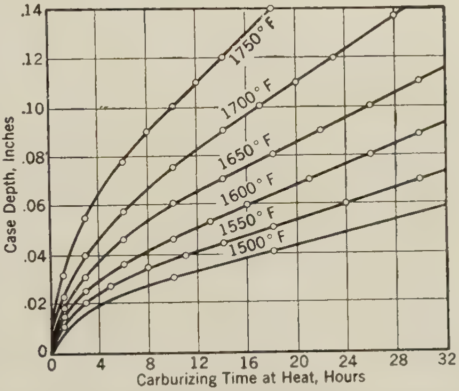
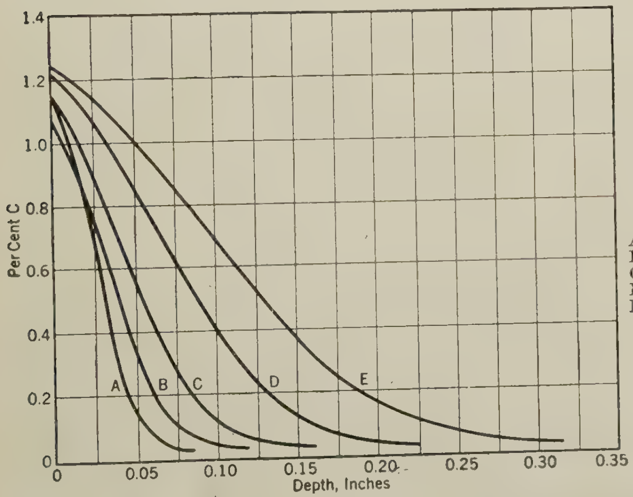


FIG. 6.—Carburizing rates, using natural gas. (Schlumpf, cited by Williams.[4])



	Temp. deg.	Time hr.	Weight increase g.
A	1739	5	1.70
B	1747	10	2.00
C	1748	20	2.75
D	1748	40	3.85
E	1745	80	5.05

FIG. 7.—Depth of C penetration in gas carburizing, with CO saturated with benzene vapor at 60°. (Bramley and Lawton,[3] as cited by Grossmann.[4])

fusion. The carbide of an alloying element may be less soluble than iron carbide. This is the case with V.

The results of different investigations on the variables affecting case depth are likely to be somewhat inconsistent, partly because of differences in methods of measuring case depth, and partly because different lots of steel of similar composition actually vary in rate of C absorption and diffusion, grain size being one variable. The steel itself may vary both as to maximum C taken up and as to rate of diffusion. Cr and Mn steels show more rapid, and the Ni steels less rapid, carburization than plain C, Fig. 5, while in the same composition, as shown in Fig. 1, coarse-grained steels carburize slightly more rapidly than fine-grained ones. The effect of alloys will be discussed in Chap. 14, Vol II. For the present we need consider only C steels. From the figures it is seen that 10 hr. at 1700° will produce a case depth varying from 0.04 in. to 0.10 in. deep, depending on those various factors.

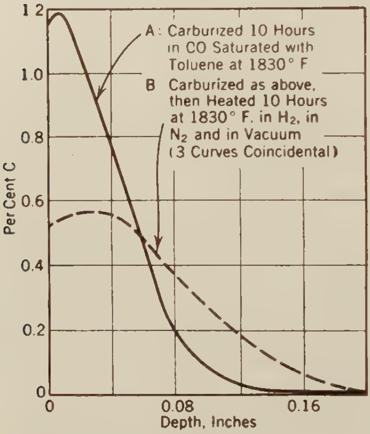


FIG. 8.—Diffusion of carburized case. (Bramley, cited by Williams.[4])

Measurement of Case Depth. Difficulties in comparing data as to depth of carburizing result from various investigators using different methods of measurement. The American Society for Metals [6] suggests that the total case be measured by fracturing or sectioning a quenched piece, etching in dilute nitric acid, or heat tinting on a hot plate. The high-C rim will etch, or blue on heating, before the low-C core. With the line of demarcation thus made more clear, depth may be measured under a low-power microscope.

The *effective* case can be determined by taper grinding and determining the hardness by Rockwell cone, or Vickers diamond. Fifty-eight Rockwell C or 650 Vickers would usually be considered full hardening, though 60 Rockwell C or 700 Vickers might be required. The exact method of depth measurement used should always be stated.

Distribution of Carbon within the Case. The surface needs a certain level of hardness and wear resistance, but the need for these properties is accompanied by an equal need for adherence of the case. If the surface is too brittle, or is underlaid by a brittle layer, the case may spall off. Hypereutectoid steel is more brittle than eutectoid steel, and it is especially brittle when the cementite is in a network or in spines, as in Fig. 6*b*, Chap. 4. On slow cooling of high-C austenite there is a strong tendency for separation of the excess carbides in these embrittling forms. On the other hand, if the carbide is distributed in something approaching a spheroidized form, the hypereutectoid carbide does not cause embrittlement, and adds to wear resistance. Unless there is opportunity for diffusion there may be a sharp line of demarcation between the case and the core; the case may be just a shell that can readily be peeled off. Such a condition is of course not met if the case is not hypereutectoid; and it would suffice to make a tough, adherent case, to limit its exterior C content to the eutectoid composition, which can be done by using mild carburizers at low temperatures, and is done when case toughness is more important than its wear resistance. But the actual problem is more commonly to produce a case of 1.10-1.20% C at the outside, for wear resistance. The Acm temperature for 1.10% C is at about 1600°, for 1.20% at about 1675° (see Fig. 6, Chap. 2). If the C concentration rises above the Acm percentage for a given carburizing temperature, or if with a C content close to that of Acm at that temperature, the temperature is allowed to drop, carbide will be thrown out. On raising the temperature it will not very readily be dissolved again.

The Carbon Gradient. Great stress is laid on avoiding a sharp line of demarcation between case and core, that is, the C content, depth gradient curve should not be steep. Such a curve as the upper one in Fig. 10 is to be avoided, for the sharp drop in C content at about the eutectoid composition makes such a case liable to peeling or chipping on that line, a phenomenon sometimes called "enfoliation" or "exfoliation." The gradient in the lower curve of Fig. 10 is obtained with a milder carburizer and at a lower temperature.

The plane of weakness in the upper curve of Fig. 10 is not necessarily present just because the surface C is high. This plane of weakness can be accentuated or obliterated by variations in cooling. Curve *b* in Fig. 11 is for a case produced with pure ethylene, a strong carburizer, but the steel was quenched from the carburizing temperature. The gradient so secured is good. On cooling steel so carburized in an atmosphere that was decarburizing, C was removed from the surface and diffusion was taking place in both directions, accounting for the hump in the upper curve.

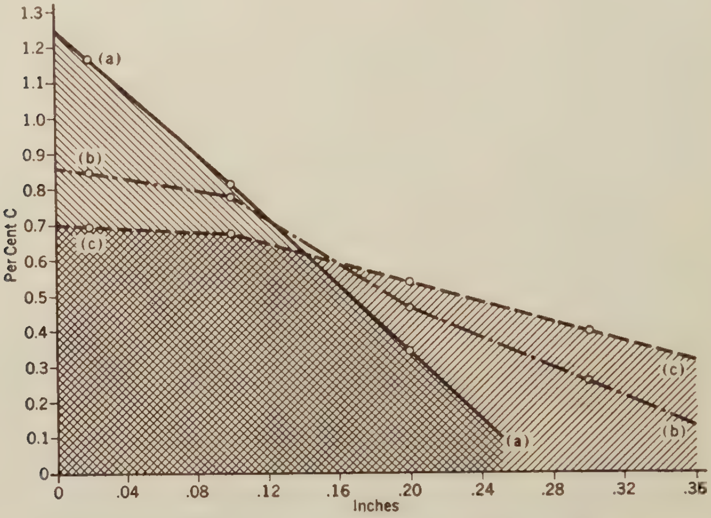


FIG. 9.—Diffusion. (a) Carbon gradient after 10 hr. in solid carburizer at 2010°. (b) Same, then held 5 hr. at 2010° in CO. (c) Same as (b) but held 5 hr. more at 2010° in CO. (Giolitti.[5])

Similarly, if carburizing is stopped and a diffusion period allowed, the C gradient may be flattened, as in Figs. 8 and 9, i.e., the total

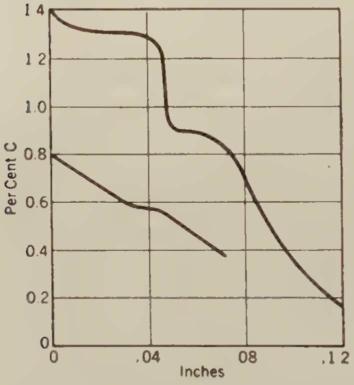


FIG. 10.—Carbon gradients. Upper curve, carburized 4 hr. at 1920° with ethylene. Lower curve, carburized 4 hr. at 1830° with 97% CO, 3% ethylene. (Giolitti.[5])

depth of case may be increased while the C is lowered. Such diffusion minimizes the danger of getting a sharp line of demarcation and a "shelly" case.

[illegible]

The two phenomena, C absorption at the surface and C diffusion, are differently affected by temperature. Figure 12 after Houdremont and Schrader[7] shows that as the surface C reaches saturation—the highest C content the particular type of steel will take up at that temperature from the particular carburizing atmosphere used—diffusion, i.e., depth of case, pulls ahead of the carbon pick-up, and the gradient flattens.

Control of the Gradient. Cook-book recipes for getting a good gradation between core and case are of little avail. Knowing the depth of case needed and its proper structure and surface hardness, and the desired structure and properties of the core, the carburizing conditions and the heat treatment are juggled to fit the particular lot of steel used. Pilot pieces are put through along with the work, file tested, and broken to show the nature of the case and the core and the adhesion between them.

The C content is controlled by the carburizing temperatures (higher temperature making for higher carbon save at the very surface) and by the "strength" or "mildness" of the carburizer. Low-C cases are produced by using little energizer and a large proportion of spent compound in solid carburizing. High-C cases are produced by the opposites or by introducing oil or other hydrocarbon, such as methane, CH_4, as well as compound. In gas carburizing, high temperatures and a high hydrocarbon concentration make for high C. By using some solid carburizing compound or by dilution of the hydrocarbon gas with flue gases or the controlled products of partial combustion of the hydrocarbon or some other hydrocarbon gas, CO and CO_2 are introduced, the CO_2 reacting with the soot deposited by the hydrocarbon, so that both the $CH_4 : H_2$ and the $CO\text{-}CO_2$ carburizing mechanisms are utilized. The first gives high C, the second low, so any intermediate condition may be secured.

Mechanism of Carburization. Having followed the general behavior of any carburizing agent, we may now examine the different ones in more detail. In contact with Fe, at the carburizing temperatures, the carburizing gases react according to the following equations, assuming the formation of Fe_3C.

$$2CO + 3Fe \rightleftharpoons Fe_3C + CO_2$$
$$CH_4 + 3Fe \rightleftharpoons Fe_3C + 2H_2$$

Both these reactions are reversible. By starting with CO or CH_4, C can be added to the surface of Fe; or, starting with CO_2 or H_2, C (as C, or from Fe_3C) can be removed from the surface (that is, from the austenite in which it is dissolved). In order to keep on adding C,

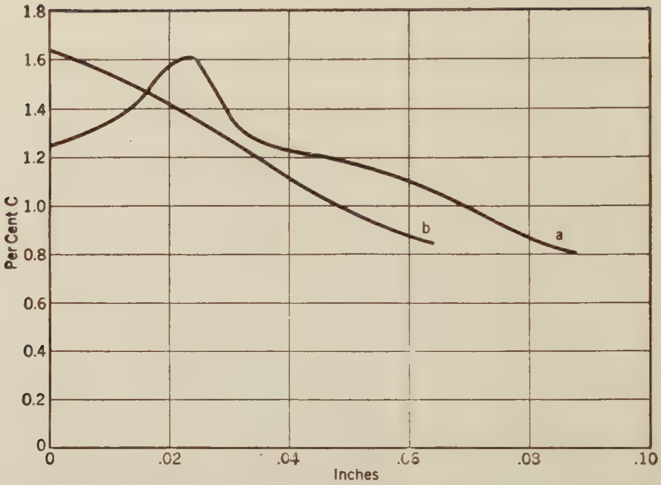


FIG. 11.—Carbon gradients. Curve (a), carburized 4 hr. in ethylene at 1830°, slow cooled in a decarburizing atmosphere. Curve (b), carburized 4 hr. in ethylene at 1830° and quenched. (Giolitti.[5])

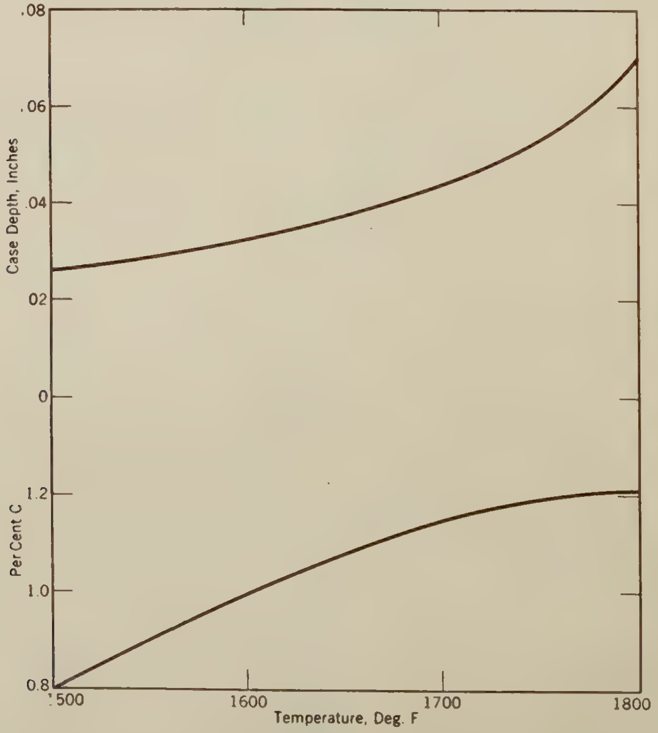
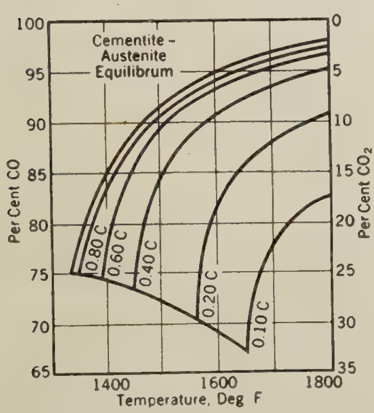


FIG. 12.—Relation of case depth to surface C content. Steel held 10 hr. in solid carburizer at temperatures shown. (After Houdremont and Schrader.[7])

we must remove the CO_2 or the H_2, as the case may be, and replace it by more CO or CH_4, else the action will stop. For any given temperature and any given C content of the steel surface which they touch, there is some definite ratio of CO : CO_2 (Fig. 13) or $CH_4 : H_2$ at which the carburizing and decarburizing tendencies are just balanced. At 1750° about 98 parts of CO to 2 parts CO_2 are required to keep the gas mixture slightly carburizing to high-C steels, whereas less than 1 part of CH_4 to 99 parts of H_2 is required. (See Figs. 9 and 11, Chap. 12.) Starting with pure CO only a little CO_2 can be allowed to form without having carburizing stop; but starting with pure CH_4 a very large amount of H_2 can form without stopping it.



Schenck's curves for equilibrium of CO, CO_2 mixtures with steels of different C content. (Quoted by Webber.[8])

FIG. 13.—This figure shows, for example, that, when sufficient time has been allowed for equilibrium, a 90 CO, 10% CO_2 gas at 1500° will carburize a low-C steel up to 0.80% C, but no higher. At 1600° this gas would carburize it only to 0.40% C. With a carburizing temperature of 1700°, to produce a hypereutectoid case, the gas needs to be 97% CO or more, with 3% CO_2 or less. In box carburizing, a reactive form of C, and an "energizer" are utilized to ensure a sufficiently high ratio of CO to CO_2.

Sources of Carbon, Solid Carburizers. In carburizing with CO, then, great pains have to be taken to eliminate the CO_2 as fast as it is formed. The best way to do this is to change it into CO, as is done in commercial practice, by supplying solid carbon which, at these temperatures, reacts as follows.

$$C + CO_2 \rightleftharpoons 2CO$$

This reaction also is reversible and will itself come to a stop when the CO: CO_2 ratio becomes high enough. Fortunately, the CO content of the gas in equilibrium with C is enough higher than that in equilibrium with Fe carbide (in the amounts present in high-C austenite) so that the presence of C makes the gas rich enough in CO to maintain the carburizing action.

Grossmann[9] has shown that carburized steels packed in carburizing compound will decarburize if held at 1250°. That is, the CO : CO_2 ratio that exists at that temperature in equilibrium with C is so low that it is actually decarburizing to eutectoid and hypereutectoid compositions. (Compare Figs. 10 and 11, Chap. 12.) The converse of this observation is the fact that, other things being equal, the higher the temperature used, the higher the C content of the case.

Mechanism of Box Carburizing. We can picture the action as follows: two molecules of CO hit the steel surface and give up an atom of C to the steel, forming one molecule of CO_2.

The CO_2 molecule wanders over to some nearby C and picks up one atom of C to re-form another pair of CO molecules, ready to repeat the performance. The solid carburizer does not need to touch the steel since the action is all by means of gas, but it is best to have the carburizer close to the steel in order that the gas may easily adjust itself in this way.

Energizers. The richness of the CO may be further enhanced by mixing barium or sodium carbonate with the C. These compounds, at these temperatures, slowly react with C as follows:

$$BaCO_3 + C \rightleftharpoons BaO + 2CO$$

or

$$Na_2CO_3 + C \rightleftharpoons Na_2O + 2CO$$

The Na_2O or BaO tends to react with the CO_2 (present from the reaction of CO with steel) as follows:

$$BaO + CO_2 \rightleftharpoons BaCO_3$$
$$Na_2O + CO_2 \rightleftharpoons Na_2CO_3$$

Thus the presence of these oxides tends to take up the CO_2 and so to give a gas lower in CO_2, and thus higher in CO, than would be the case without them. These compounds are called "energizers" because they produce a more energetic carburization.

Another explanation for the effect of the energizer is the reversible formation and absorption of CO_2 according to the reactions

$$CaCO_3 \rightleftharpoons CaO + CO_2$$
$$BaCO_3 \rightleftharpoons BaO + CO_2,$$

by which at low temperatures CaO or BaO absorb CO_2 and at high temperatures give it up. The added amount of CO_2, reacting with C, forms more CO. This augumented supply of CO is postulated to be formed more slowly and steadily at carburizing temperatures, from $BaCO_3$, and too early with $CaCO_3$, thus accounting for the better performance of $BaCO_3$. This is the explanation most commonly given, but the explanation based on the catalytic effect of the oxide, i.e., a displacement of the $CO : CO_2$ equilibrium, appears more satisfying.

It is possible to carburize in a closed container with only the use of hardwood charcoal and the air that was in the container when it was closed. The oxygen of the air unites with the charcoal to give the $CO:CO_2$ ratio obtaining at the temperature of operation. This ratio is just great enough to be carburizing to steel under 0.90-1.00% C and neutral or decarburizing to steel of higher C contents, so the maximum C content in the surface layer is only of approximately the eutectoid composition.

With the energizer present, and hence a high ratio of CO to CO_2, the action will go further, so that cases of, say, 1.20% C or even higher can be obtained by letting the action proceed long enough and at high enough temperature.

Carburizing Primarily a Surface Action. It is sometimes considered that, instead of being a purely surface reaction, the carburizing action involves the diffusion of CO below the surface of the steel before it breaks up. If this did occur, the CO_2 formed as the two molecules of CO give up one atom of C, would oxidize the Fe that is right there instead of being able to escape and wander away to find an atom of C. As previously stated, very tiny amounts of O_2 are taken up by the steel in box carburizing, but this seems to be a surface reaction only. If gas diffusion into the metal does occur, it must be only to a very slight extent. The great bulk of the action is doubtless at the surface. Austin [10] points out that in decarburization there is no experimental justification for the assumption that gas diffuses into the steel and then diffuses out, carrying C with it. The action is wholly at the surface and the loss of C is due to diffusion of C to the surface when it is removed. A similar point of view in regard to carburization seems justified, though Austin himself appears not to hold it for carburization.

Smithells [11] considers that a compound gas like CO cannot diffuse as such, though Fe, while C and O, by themselves, can. The theory of diffusion of CO as such, beneath the surface in the carburizing process, is not borne out by direct experiment or by analogy. It seems certain that CO acts only at the steel surface and is there decomposed. The nascent C is readily absorbed and diffused inside the steel, while the nascent O is less readily absorbed and diffused. The possibility of absorption of nascent O, however, differentiates carburizing with CO from carburizing with a hydrocarbon gas.

The Introduction of Oxygen. Obviously, if the CO_2 molecule meets Fe before it meets C, the reversible reaction

$$CO_2 + Fe \rightleftharpoons CO + FeO$$

comes into play. In the body of the gas, CO is in great excess. The reaction can only momentarily go to the right, for diffusion of CO from the body of the gas will force the reaction to the left, i.e., the FeO that has been momentarily formed will be reduced. In the interim between formation and reduction of FeO, the FeO may diffuse into the steel. As Grossmann[9] has shown, there may be a slight pick-up of O_2 in box carburizing. This slight influx of O_2 is thought to account for the "abnormality" of the case in carburized steels containing traces of Al, the incoming oxygen reacts with the Al to form Al_2O_3, and the presence of very finely divided Al_2O_3 particles interferes with the diffusion of C to the grain boundaries.[12, 13] The explanation is consistent with the fact that steels that are abnormal when box carburized, are normal when gas carburized with a hydrocarbon gas free from oxygen-containing gas. (Compare Fig. 1, Chap. 9.) Derge, Kommel, and Mehl[14] show that Al_2O_3 particles present in the steel before carburization do not produce abnormality, nor does the presence of Al in solid solution (in the absence of oxygen in the carburizer). The O_2 is introduced in pack carburizing. Bain[12] has shown that very pure Fe, oxygen-free, and carburized in absence of O_2, is normal, while high-purity Fe contaminated by O_2 may show abnormality even when carburized in oxygen-free hydrocarbons, so the presence of Al is not essential to abnormality. However, commercial steels used for carburizing contain so little O_2 in solid solution that they are ordinarily normal if carburized without addition of more O_2.

Varieties of Solid Carburizers. Quite a variety of carbonaceous materials may be used for carburizing. The usual solid carburizers are made up essentially of hard-wood charcoal and barium carbonate ($BaCO_3$), the pulverized mixture being bonded with molasses or oil as binder and formed into pellets. Pellets are used to keep the $BaCO_3$ in close contact with the charcoal and avoid its settling out, to avoid dustiness when the charge is dumped, and to make the packing offer less resistance to the flow of heat into the charge. Modern practice in compounding carburizers is discussed by Widrig.[15]

Oil may be added to the solid carburizer when the box is packed, with the idea of displacing the air in the box by oil vapor before the work gets up to carburizing temperature, so that preliminary scaling may be avoided.

Charred leather or charred bone were old-time favorite types of C, since their ash probably has energizing action and some hydrocarbons may be evolved on further heating of incompletely charred organic matter of this type. The charred bone compounds are less active carburizers than the energized charcoal type, Fig. 3, but retain their effectiveness in use without addition of fresh material. They do not shrink so much with use. They are especially applicable when deep cases with low C at the surface (eutectoid or only slightly hypereutectoid) are desired, and when long heating times are employed. They are "mild" carburizers.

Where a very rapid box carburizing, to produce a very thin, but very hard surface layer, is desired, sodium or potassium ferrocyanide may be added to the carburizer in order to give a skin hardened both by C and by Ni, but this expedient is very little used, most such cases being produced by liquid bath surface hardening.

Pulverized coke is often used as part of the C, but is not as reactive as charcoal, so it is generally considered as a diluent. Without addition of energizer, ordinary coke is almost without effect as a carburizer, and sugar charcoal is likewise practically inert, while hard-wood charcoal alone will act as a mild carburizer. The explanation for this doubtless lies in the fact that the ash of hard-wood charcoal contains sodium carbonate, Na_2CO_3, which is an energizer.

Many solid carburizers contain Na_2CO_3 as the added energizer, but as the Na_2O formed is more destructive to the carburizing box than BaO, the latter is considered preferable. $CaCO_3$ is also used, but not usually by itself. Many authorities consider that $CaCO_3$ is either inert or at least a much less active energizer than $BaCO_3$ or Na_2CO_3.

Reuse of Spent Carburizer. Spent carburizer may be allowed to stand in the air for some time, say overnight, in order to be "regenerated" before reuse. The usual explanation is that the BaO must be allowed to take up CO_2 from the air and regenerate $BaCO_3$. This explanation is not satisfying, for fresh carburizer made up with BaO works just as well as that made with $BaCO_3$. Charcoal readily absorbs air upon its surface and gives it off slowly on heating, so that an equally good explanation would depend on the replacement of the adsorbed air that had been driven off in use.

Obviously, there must be enough O_2 in the pot at carburizing temperature to make CO and CO_2. If the new carburizer added contains oil or molasses as a binder, the cracking of these organic materials will tend to drive the air initially present in the box out through the clay luting, so that O_2 in some form that will not be displaced from the box before the carburizing temperature is reached must be supplied. When the carburizer is made up from charcoal and BaO, the O_2 that forms CO and CO_2 must largely be that adsorbed upon the charcoal. If the O_2 supplied had to come from the decomposition of carbonates, there would be more relation between the action of the various carbonates and their decomposition temperatures than there is. The actual relation appears to relate far more closely to the catalytic action of the Na_2O, BaO, etc.

If other means were taken to ensure the presence of the necessary small amount of oxygen, it should be possible to reuse "spent" carburizer without a period of storage.

Amount of Energizer. The amount of energizer added to the pellets varies from 5 to 20%. Excess energizer over perhaps 2 or 3% does not make for appreciably higher C content of the case, or for more rapid carburizing, but since, with use, the energizer may sinter down with the ash of the C and become less effective, an excess is added at the start so that the carburizer may be reused. A mixture of some proportion of used and some of new carburizer is used in practice, the proportions varying with conditions and being usually established on a cut-and-try basis. According to Davis[16] the only new carburizer added in modern practice is that required to make up losses from dusting and burning.

It has even been suggested that, when a low-C surface layer is desired, and when working at high operating temperatures to allow rapid diffusion, an "anti-catalyser" be added. Mahin and Spencer[17] claim that a small amount of ferrosilicon has such an anti-catalytic effect.

Boxes and Their Packing. It is necessary that the access of carburizing gas to all sides of the work be uniform. Work will carburize even if not submerged in the carburizer; but the $CO:CO_2$ equilibrium will not be so readily established at a distance from the charcoal, and the carburizing action will not be so uniform. Hence the boxes are packed so that, as the carburizer shrinks and settles, it will still envelop the charge. Tightly fitted covers, well luted on with clay, are used to avoid waste of carburizer by combustion with air or flue gas. Boxes and carburizer are dead weight to be heated up, so the greater the amount of work to the weight of carburizer and box the better. Heavy cast boxes are being replaced by thinner boxes, often ribbed for stability, made of heat-resisting alloys to prevent scaling of the outside of the box.

Size of Carburizing Box. The inside of even a small box takes quite a while to come up to the temperature of the furnace; and, if a large box is used, the material in the center lags behind the indicated furnace temperature several hours, i.e., several hundred degrees. The greater the size of the box, the larger will be this error and the greater the actual difference in the thickness of case taken on by steel near the sides of the box as compared with that near the center of the box. No manipulation of the furnace can change this effect; it can be remedied only by altering the dimensions of the box itself.

The box should not be larger than is absolutely necessary, even where large quantities are to be carburized in it. It should be narrow in at least one dimension so that the heat may have a chance to penetrate quickly, at least from two sides, and reach all the contents at about the same time. Further, the boxes should not be made too deep in proportion to their other dimensions, as this makes packing the parts into them more difficult. The design of the box may, in special cases, follow the outline of the piece to be carburized, about ½ in. all around being allowed for clearance and packing, so that the surfaces may be heated uniformly and carburized alike. Several such containers may be mounted on a stand, the whole being handled as a unit. Roush and Dames [18] have given a good discussion of containers.

If the work is to be cooled in the box it is important that the box is not so large that the work cools too slowly, for slow cooling makes the hypereutectoid C tend to assume the network form; rapid cooling avoids it. Hence the closer box cooling approaches to normalizing, the better, though a balance has to be struck when the work has to be machined after carburizing and the case therefore is required in the annealed condition. Thin-walled boxes and the minimum of space for compound favor both rapid cooling and full efficiency.

Packing the Boxes. The method of packing should be such as will ensure as nearly as possible an even heating and a uniform carburizing of all pieces in the same box. The method of packing necessarily must vary with the type of article to be handled. For example, long, slender pieces should be packed vertically, so that they will be held in position by the carburizing material and cannot sag under the influence of the high temperatures. Again, gears and similar pieces may be most suitably packed in tubes, so that the same amount of carburizing material and the same degree and length of heating may influence all parts of the periphery in equal proportion. Should a narrow or low box not be available in connection with small work, carburizing compound that has been used before may be put next to the sides of the box while new compound is placed in the center. In this way the difference in carburizing which might result from the different temperatures in various parts of the box may be offset.

The first step in the general operation of packing is to cover the bottom of the box with the compound to a depth of 1½ to 2 in., tamping it solidly into place. The parts to be carburized are then placed firmly upon this bed so that the compound and work are in close contact with each other. The pieces should in no case touch the sides of the box, but should be placed about ½ to 1 in. away from it. The articles should be separated from each other by about ½ in. If the articles should touch one another at any point, soft spots would result. Non-uniformity of case may also result if there is not sufficient carburizing material in the box.

After the first layer of work has been placed in the box, it is entirely covered with the carburizing compound. This should be packed and tamped down around and over the pieces so as to have the particles of compound in close contact with the steel, but yet not so tightly as to prevent the free circulation of the carburizing gases which are generated during the heating process. When the first layer has thus been suitably packed and covered, the same procedure is repeated until the box is nearly filled. The layers may be separated by cardboard. The point to be kept in mind is that each and every piece should be surrounded on all sides by a suitable amount of the carburizing compound.

At least 2 in. of the compound should form the top blanket over the last layer of work. Some shops adopt the following practice, with the aim of further preventing the escape of the gases: about 2 in. from the top of the box, sheet strips about $\frac{1}{16}$ in. thick are laid over the last layer of the carburizing material and these, in turn, are covered with about 1 in. of powdered charcoal. When the box is finally packed, the cover is placed on the box and the edges are carefully sealed with fire-clay or asbestos cement. The box is now ready for the heating operation.

The Heating. The two important points are: the heating, at least up to 1300° should be gradual; the heating beyond this temperature should be uniform over all parts of the carburizing box. An energetic liberation of gases commences very strongly at temperatures somewhat under 1300° for the majority of solid carburizing compounds, and it is advisable to diminish this as much as possible in order to obtain a more gradual carburization. Furthermore, it gives more opportunity for the steel to heat uniformly and avoid warping. The second point is self-evident: non-uniformity of heating must necessarily result in non-uniformity of product. The boxes can be heated in batch furnaces, or in continuous, conveyor type furnaces, sometimes made on the counter current principle with two rows of boxes moving in opposite directions so that the emerging hot boxes give up some of their heat to incoming cold boxes, when the work is to be cooled in the box. If it is to be quenched from the box, the counter current principle is not employed.

The temperature can be dropped toward the end of the run in batch carburizing, or a lower temperature zone can be provided in a continuous furnace, to allow a period of diffusion without continued pick-up of C at the surface.

Gas Carburizing Furnaces. In gas carburizing it is essential that the gas composition meeting the work shall be uniform. If gas is passed in at one end of a charge, it will be somewhat depleted of its carburizing constituents by the first work it reaches, so care is taken to produce turbulence of the gas and to avoid channeling in a packed charge of small parts whereby the gas would seek the open channels and avoid the more tightly packed portions.

Hence the charge may be tumbled, in a slowly rotating drum when the work will stand agitation without damage, the gas may be led in at a number of places, or the gas may be circulated by a fan. In the rotating drum type of furnace, some solid carburizer may be added, as well as a hydrocarbon gas.

In any case, the charge and carburizing gas must be held in a muffle-like shell or container to protect against air or flue gases. The container can be externally heated, usually either by gas or electricity.

When the work will not stand tumbling, the gas-carburizing furnace has an outer shell within which is a combustion space (or electric heating elements), the container, and within the container a basket into which the work is loaded, and either a number of inlet pipes and an outlet for the carburizing gas, or one inlet with a fan either above or below the basket to circulate the gas. Container, basket, and fan are all of heat-resisting alloys. Figures 8 and 9, Chap. 15, show an electric carburizing furnace. Figure 14 shows batch and continuous, gas-filled, gas-carburizing furnaces.

McQuaid[19] cites an electric carburizing furnace which has the resistors inside the retort, radiating directly to the work. Straight natural gas is led in and allowed to crack for 3 min., then the gas supply is cut off for a 27-min. diffusion period, and the process repeated. By this technique, short circuiting of the resistors by too thick a coating of soot is avoided. The thermal efficiency of a furnace of this type would be high. The resistors would have to be of material that does not itself carburize readily and deteriorate.

Gases for Carburizing. Given suitable equipment, a wide variety of carburizing gases may be used. Manufactured, or "city" gas, varies

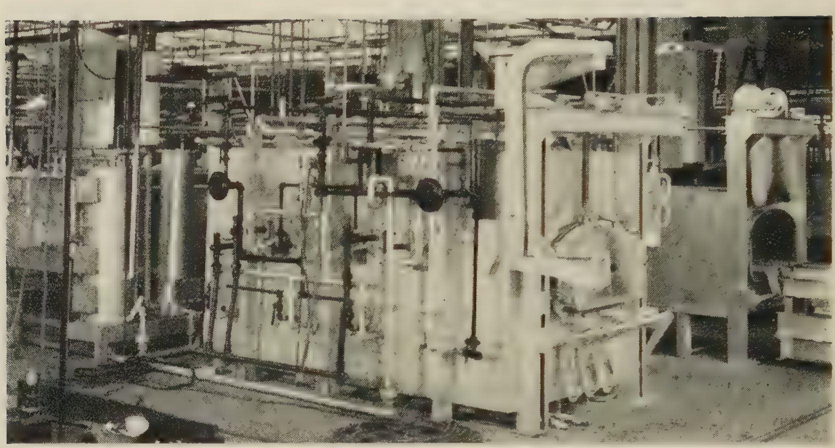


Fig. 14a.—Batch type, gas-fired, gas carburizing furnace at Packard Motor Car Co. The gas flames heat a muffle. The work is loaded on trays. The trays are pushed through the furnace, progressively passing through three different temperature zones. Gas locks are used to prevent loss of carburizing gas. The carburizing gas is prepared in the equipment at the left. (Courtesy Surface Combustion Co.)

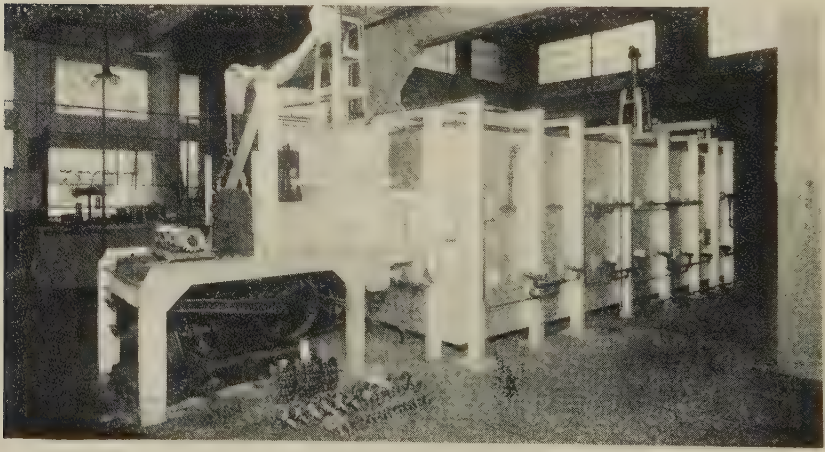


Fig. 14b.—Continuous gas-fired, gas carburizing furnace at Reed Roller Bit Co. (Courtesy Surface Combustion Co.)

in composition according to the way it is made and how it is enriched to meet the local regulations as to B.t.u. It usually contains large amounts of hydrogen, H_2 and methane, CH_4, with lesser, and often varying, amounts of higher hydrocarbons, unsaturated hydrocarbons, and illuminants. These more complex hydrocarbons tend to break down into CH_4 and H_2 at carburizing temperatures, but in so doing may deposit undesired soot upon the work.

Many different hydrocarbons will serve for carburizing. Besides methane, ethane, butane, propane, ethylene, turpentine, benzol, pine oil, mineral oil, etc., may be used. At carburizing temperatures they all tend to break down into soot, CH_4 and H_2. While their action may vary during the moment of breakdown, after the soot is deposited the atmosphere is primarily one of $CH_4 + H_2$. Hence they all act much alike,. and give a CH_4:H_2 ratio high enough in CH_4 for carburizing.

Methane itself has a tendency for thermal decomposition, i.e., cracking, and the deposition of soot, until this action is opposed by the presence of a high enough hydrogen content. This cracking reaction will go on in the absence of Fe. But even when equilibrium of the CH_4:H_2 ratio in presence of C has been established, contact of the CH_4 of that mixture with Fe produces more cracking, liberation of H_2, and the formation of nascent C which is absorbed by the austenite just as is the nascent C from CO in the case of solid carburizing. Admission of some oxidizing gas, such as flue gas, to react with soot, is utilized to offset this. The oxidizing gas may be mixed with the carburized gas, or be admitted at intervals.

Surge Methods *vs.* Steady Flow. McQuaid[20] criticizes the intermittent surge method and the method of introducing oxidizing gas along with a rich hydrocarbon to react with soot deposited from the latter. Admitting that both work in practice, he considers them more difficult to control than to maintain a uniform, turbulent flow of gas of regulated stability and dilution that will give up active C at the desired rate without strong tendency for the accumulation of soot. A more stable hydrocarbon, such as the methane of natural gas, diluted with city gas, is, therefore, preferred by McQuaid to propane or butane which are less stable and so have a stronger sooting tendency.

Gable and Rowland[21] remark that practically all commercial installations utilize a steady flow of diluted gas rather than the intermittent surge method.

The presence of O_2 or water vapor with a hydrocarbon gas introduces complicating factors, since O_2 will react with C, Fe, and Fe_3C, and water vapor with these and with CO_2 as well. Hence the diluent might better contain only CO and CO_2, without free O_2, and would preferably be chilled to a low and constant content of H_2O. Some gases work better with a small amount of moisture present. If the moisture content varies, as is usually the case with city gas, it may be necessary to install equipment that will first dry the gas, irrespective of the fluctuations, and then humidify it back to just the desired extent.

In the use of a solid carburizer, to give CO, there is an excess of C present as charcoal and the operation is carried out in a closed chamber, while in carburizing with hydrocarbon gases, the used gas is run out and fresh gas admitted, to provide the needed C. Gas composition and rate of gas flow, in respect to area of steel surface to be carburized, therefore, are factors that may be varied.

Effect of Containers. An additional complication is that the material of the carburizing pot of the trays or baskets in which the stock is held in gas carburizing may exert a catalytic effect. Cowan [22] remarks that the proximity of 35 Ni, 15% Cr trays to the work slows down the rate of gas carburizing. If the trays are preoxidized before use, the rate of carburizing is not decreased.

Hydrocarbons Available for Gas Carburizing. Where it is available, and it is now very widely available, natural gas is a favorite carburizing gas. In any given location its composition is very constant, though different natural gases will differ, some running almost pure CH_4, while others contain rather large amounts of higher hydrocarbons.

The higher hydrocarbons, propane and butane, are now available in liquid form, being shipped in tanks or tank cars under pressure. The liquids gasify when the pressure is released. These commercial gases can be had quite pure and of quite constant composition at such prices that they can be considered as fuel gases where natural gas is not available. Hydrocarbon gases can be made on the spot by dripping mineral oil or vegetable oil (e.g., pine oil) into the carburizing container, in which case it is necessary to be sure that coke from the cracking of the oil does not collect upon the work, but that the oil drips directly onto a shield. Or the oil can be cracked in a separately heated unit, the coke and excess soot deposited there, and the cracked gas passed on to the carburizing unit. (See Fig. 25, Chap. 12.)

Carburizing by a Combination of CO and Hydrocarbons. The rich hydrocarbon gases, natural gas, propane, especially butane and cracked oil vapor, are richer than necessary and may form troublesome soot deposits. They may be diluted by partial combustion with a deficiency of air, when the resultant gas will contain CO, CO_2, unburnt CH_4, H_2, and water vapor. The water vapor may be mostly removed by chilling the gas. It is difficult to control partial combustion so that the gas produced will be sufficiently carburizing, but unburnt gas can be mixed with the partly burnt gas to enrich it.

When an unburned hydrocarbon is used that deposits too much soot, it may be diluted with flue gas containing CO_2, or a little air may be admitted. In actual gas-carburizing practice straight hydrocarbon gases, free from CO, are seldom used, as the balance between a sufficiently active gas and one that will deposit too much soot is hard to maintain. Davis[23] reports that a hydrocarbon gas diluted with flue gas carburizes faster than the hydrocarbon alone, the increase being from a case depth of 0.04 in. in 5 hr. without flue gas to that depth in 4 hr. with it.

Obviously the CO and CH_4 methods of carburization can be, and often are, combined. Oil is sometimes added to the charcoal-$BaCO_3$ mixture, to break down into CH_4 and H_2; and, conversely when city gas, natural gas, or other hydrocarbon gas is used, some solid carburizer may be put in with the charge to supply some CO.

Thus there are many ways of producing a carburizing gas and the one chosen will depend on the cost of the gases or oils available and on the constancy of composition of those available. The effect of a given gas mixture on the particular steel being used, at the time and temperature chosen, can be so readily checked by examining the case produced that the details of control are worked out locally rather than by adhering to any general "recommended practice" set up by outsiders.

As Brearley[24] put it in regard to box carburizing, "no man's word is law in an experimental craft like case hardening"; and this is equally true in gas carburizing.

Armor Plate. Quite different from the small parts and the relatively thin cases with which carburizing processes are ordinarily concerned is the carburizing or "Harveyizing" of heavy armor plate for battleships. In this process huge pieces of plate, sometimes 12 ft. × 10 ft. × 14 in. thick, are placed in a huge furnace with carburizing compound on what is to be the outer face of the plate, the back being shielded from carburizing, and heated for as long as 3 weeks. Krupp gas carburized some 2 weeks at 1740°, cooled to around 1600°, oil quenched, drew at 1170–1290°, after which the plates were bent to shape, then differentially heated, the carburized face being heated to 1560–1650° while the back face was kept at 840–1020° and then water quenched by pressure sprays from both sides, in order to produce a hard, high-C surface upon a core itself of maximum strength and toughness. According to Ehrensberger[25] the average C content of the first 0.2 in. is 1.56%, of the second 1.08%, and of the third 0.64%.

The steel employed is usually a highly alloyed, air-hardening steel, such as 0.35–0.40 C, 3.75 Ni, 2% Cr with a small Al addition. The processes are carried out at only a few armament works, and their details are seldom discussed in technical literature.[26] Fortunately, the average heat treater does not have to deal with so difficult a problem as securing the proper gradation of hardness and toughness in pieces of the size of armor plate.

A somewhat analogous process is used in plates to be made into mold boards for plowshares, where a high-C, wear-resistant surface and a tough case are required. Anderson[27] describes the process in which S.A.E. X1015 steel plates 8 ft. × 11 to 17 in. wide and ¼ to ⅜ in. thick are box carburized for 22 to 26 hr. at 1700°. The packing of these plates is interesting, since stacks of plates with carburizing compound between the plates are put into wooden boxes and these boxes charged into the muffle and in turn surrounded by compound. The wood chars, but serves to hold the compound in place.

Summary. To summarize, generally speaking, the thicker cases, requiring long heating periods, are more commonly produced by box carburizing, while the thinner ones are being more and more produced by gas carburizing, though either process may be used for either thick or thin cases. A case of ⅛ in. is considered thick and one of 1/16 in. is perhaps the average sought, with around 1/32 in. as the case often sought in gas carburizing. The gas process lends itself well to continuous carburizing. In either process the rate of case formation can be varied from around 0.01 in. or less to as high as 0.02 in. per hr., depending on the temperature. The lower temperatures, around 1650°, give slower case formation but are easier on the boxes or containers. Normal carburizing times are around 6 to 8 hr. after the box reaches temperature in box carburizing, and perhaps 3 or 4 hr. at temperature in gas carburizing.

By selection of carburizing compound, type and amount of energizer, and operating temperature, the composition of the surface layer can be controlled from an eutectoid composition to 1.3% C or higher.

By control of gas composition and temperature, similar gradation in surface C content can be obtained in gas carburizing, though the tendency is to produce a higher C case. In usual practice the gas-produced cases range from slightly higher than eutectoid, say 0.95% C, to as high as 1.4% C. By diffusion, after cutting off the gas supply, the initial high-C case can be lowered in C. For many purposes, in either process, a surface C content of about 1.15% is aimed at. Other things being equal, steels alloyed with elements that readily form carbides, such as Cr, Mn, and Mo, tend to have a slightly higher surface C content.

Diffusion is of primary importance in securing a tough case. A sharp line of demarcation between the hypereutectoid outer case and the low-C case makes for a brittle case, which is likely to spall off, either on quenching in heat treatment, during grinding, or in service. A steep C gradient is therefore to be avoided on most work. The carburizing agent, the time, and the temperature are, therefore, all adjusted or varied during the carburizing process in order to produce the desired thickness of case, surface C content, and C gradient. These requirements vary according to the use to which the carburized part will be put.

The rate of carburization is greatly affected by temperature; and, to secure even carburization of all parts of the charge, temperature variations of even as little as 10° in different parts of the charge should be avoided. Properly engineered gas-carburizing outfits allow better uniformity than is obtained in ordinary box carburizing, so that for the thinner cases, where variation in case depth is especially harmful, gas carburizing offers rather better control. The slower carburization and less steep C gradient at 1650° make for better quality, while for speed and low cost 1750°, or even, with steels not susceptible to grain growth, 1800° or 1850° can be used. When very thin cases, of the order of only 0.01 in., are to be produced, temperatures even lower than 1650° may be used. When the work is to be quenched direct, 1650° is ordinarily not exceeded.

An extremely useful assembly of modern information on carburizing is contained in the A.S.M. Symposium,[28] to which the reader is referred for a more extended discussion.

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CHAPTER 11

HEAT TREATMENT OF CARBURIZED STEEL

The carburizing process produces a duplex steel, with its original low-C core and a high-C case. Unless a low temperature and a weak carburizing agent have been employed, the slow-cooled case will be hypereutectoid, i.e., it will contain free cementite at the boundaries of the austenite grains existing while it was forming. This brittle network, left undisturbed, will make the case itself brittle.

Separation of Cementite. Probably the majority of carburized parts are so carburized that the surface C content is about 1.15% or about 0.30% above the eutectoid. This corresponds to about 5% excess cementite. At 1650° or above, this amount of C can exist, dissolved in austenite without the separation of the excess cementite. If the case is quenched from that temperature the excess cementite will tend to separate in small particles rather than to collect into an enveloping network, but on very slow cooling, as when box carburized steel is allowed to cool in the box, it will so collect. Ordinary carburizing steels slow cooled after carburizing will show an outline of hypereutectoid cementite about the coarse austenite grains existing at the carburizing temperature, and this network size reflects also the size of the grains of the low-C core. Since production of this network is, or was at the time the name was applied, the normal behavior, steels that show it are called "normal."

If the austenite does not coarsen at the carburizing temperature, the carbide network (in most steels of high coarsening temperature) does not completely surround the small grains and the carbide remains in coalesced patches. Steel showing this behavior is termed "abnormal." The phrase does not now connote inferiority, since for most purposes the fine-grained abnormal (e.g., Al-treated) steels are desired. When the phenomenon was first noted a certain degree of inferiority was implied because the fine-grained steels that were abnormal, having lower hardenability because of the fine grain, as explained in Chap. 9, were more susceptible to soft spots when a gas bubble happened to cling to the surface during quenching. Introduction of elements that confer depth hardening, i.e., do not require such high quenching rates to harden, or the use of more drastic quenches or better technique in quenching have minimized trouble from this source.

It is now recognized that fine grain and coalescence of cementite are not necessarily companions, for fine-grained "normal" steels, in which the excess cementite does not coalesce but outlines the fine grains, are available. Such alloying elements as V or Ti are probably required to produce steels of this type.

Effect of Grain Size on Type of Heat Treatment. The response of the steel to the tendency toward grain growth is of paramount importance in the selection of the carburizing temperature and the heat treatment of the carburized steel. Specifications for carburizing steel no longer state merely the chemical composition; they include also the grain size according to the McQuaid-Ehn test. This test, if made at the carburizing temperature that will actually be used, is logical for the selection of carburizing steels, since it shows whether the steel will or will not coarsen in the carburizing operation; and, if it does, indicates the necessity for heat treatment to correct the coarsening.

If we use steel that does not coarsen at the carburizing temperature used (even the fine-grained steels can be coarsened at very high temperatures), the core will be relatively tough on quenching direct from the carburizing temperature, not so much so as when it is given another quench from a lower temperature, but sufficiently tough for many purposes.

The saving in time and expense when quenching may be done direct from the carburizing temperature is so great that if, with the single quench, the case is not too brittle and the core sufficiently tough, this method will always be chosen. Hence fine-grained steels with the addition of ferrite-toughening and depth-hardening elements, balanced to permit the single quench, are deservedly popular. Fine grained Ni-Mo steel (4615) finds much favor for the single quench, and Cr-Mo (4120) of suitably fine grain is being similarly utilized. Chromium-Vanadium, e.g., 6115, also has a good combination of hard carbides and fine grain on a single quench. Wyzalek [1] prefers the single quench for fine-grained Ni-Cr S.A.E. 3115 steel. The lower carburizing temperatures are preferable when a single quench is to be used.

Single *vs.* Double Quench. Even with these steels, the single quench does not give optimum properties to either the core or the case, the quench is from much too high a temperature for a high-C steel like the case, and the heating time has been too long and the temperature too high for the best properties in the core. With coarse-grained steels, the core will be much too brittle and the case may not have its carbide network sufficiently broken up on the single quench. An additional quench is called a "regenerative" quench because it regenerates or refines the structure of the case.

Single Quench. The heat treatment of carburized steels must take into consideration the low-C core and the high-C case. A 0.10–0.20% C steel will normally be heated to 1600° for quenching, a eutectoid steel to around 1400°, and a 1.15% C tool steel would normally be hardened from about 1400°, but unless the carbide is already well spheroidized a preliminary treatment at about 1600°, i.e., above Acm, would be given to dissolve the carbide. In all these cases normal practice avoids heating much above the temperatures mentioned unless the steel is resistant to coarsening. Only when the steel remains fine grained, at say 1650° or higher, may it be quenched direct from such a carburizing temperature. The quench may be either in water or oil according to the case hardness and toughness desired and the hardenability of the steel, which is dependent on its alloy content and grain size. The austenitic grain size will be retained in both case and core. At some point where the C is high enough on the radius of the piece the cooling rate will produce primary troostite, and the still higher C exterior zones will quench to martensite. In alloyed steels some austenite may be retained. In the hypereutectoid zone the carbide will be well distributed; it has no time to form a network. If the grain size is small enough, both core and case will be fairly tough. This very simple treatment of direct quenching is the one used whenever feasible. We may call this treatment (1). (See Fig. 1.)

Double Quench. Suppose now that the case produced by this treatment is satisfactory, but the core is not strong enough. We can disregard the case and treat for the core, i.e., we can reheat to about 1600°, which will refine the grain of the core by passing through Ac_3, and again quench. The core will then be tougher, and the case much as it was before. Actually we are likely to want to avoid the original quench to minimize distortion and cracking, so we may slow cool from the carburizing temperature, which allows carbide network separation in the case; but, since we are going to heat above Acm, it will again go into solution, though perhaps not entirely. The treatment is at a pretty high temperature for a tool steel, which is what the high-C case approximates and, unless the steel is one that resists coarsening, the case may be too coarse. Hence for treatment (2) we shall need a steel resistant to coarsening.

Suppose on the other hand we want to make the core just as ductile as we can, by annealing. Then we slow cool from the carburizing temperature to get an annealed core. The case is annealed too, and must be reheated and quenched to harden, but we do not want to harden the core. Then we reheat only to about 1400° and quench, which leaves the core soft—treatment (3). In this case any excess carbide of the case thrown out on slow cooling is not redissolved. Thus the treatment is best applied to steels with only a eutectoid, not a hypereutectoid, case.

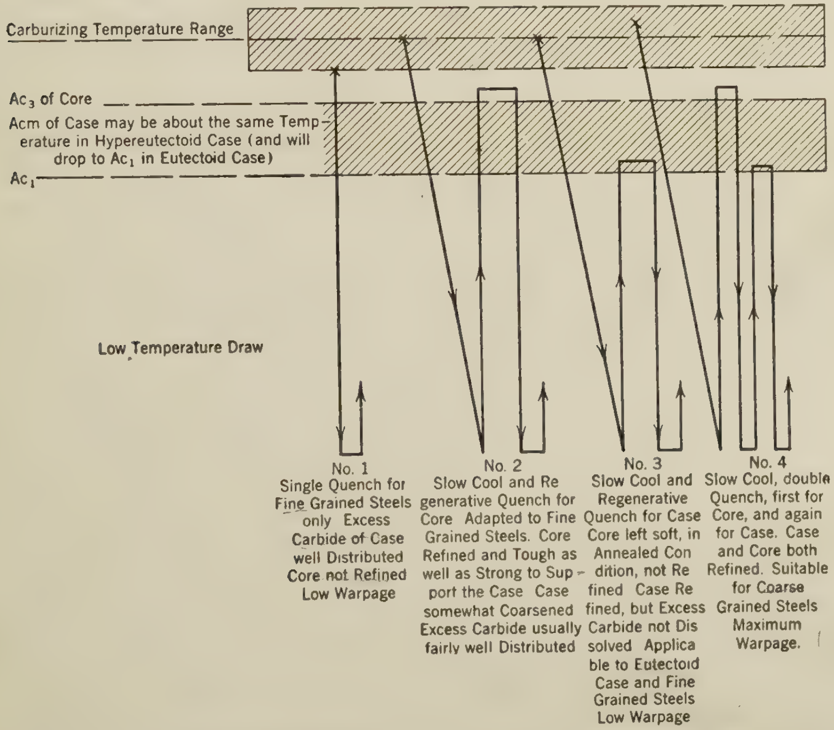
Carburizing Temperature Range
Ac_3 of Core
Acm of Case may be about the same Temperature in Hypereutectoid Case (and will drop to Ac_1 in Eutectoid Case)
Ac_1
Low Temperature Draw
No. 1
Single Quench for Fine Grained Steels only Excess Carbide of Case well Distributed Core not Refined Low Warpage
No. 2
Slow Cool and Regenerative Quench for Core Adapted to Fine Grained Steels. Core Refined and Tough as well as Strong to Support the Case Case somewhat Coarsened Excess Carbide usually fairly well Distributed
No. 3
Slow Cool and Regenerative Quench for Case Core left soft, in Annealed Condition, not Refined Case Refined, but Excess Carbide not Dissolved Applicable to Eutectoid Case and Fine Grained Steels Low Warpage
No. 4
Slow Cool, double Quench, first for Core, and again for Case. Case and Core both Refined. Suitable for Coarse Grained Steels Maximum Warpage.

FIG. 1.—Schematic representation of four methods of heat treating carburized steels. (Slightly modified from chart of Int. Nickel Co.)

Compromise Treatment. By raising the temperature, say to 1550°, within the Ac_1-Ac_3 range of the core but not above Ac_3, we get the core partly refined, a bit tougher and stronger than in (3), and some, but not all, of the free carbide of the case is dissolved. This is a compromise treatment, between (2) and (3). It is applicable to cases that are slightly hypereutectoid. In none of the above is the treatment just what would be selected as the best for a homogeneous steel of either high- or low-C content, and fine-grained steels are better adapted to, and less injured by, such treatments than coarse-grained ones.

Extended Treatment. If it is necessary to use a coarse-grained steel; or, if the maximum toughness of both case and core be required, a double quenching, first for the core and then for the case, is resorted to. The steel is slow cooled, reheated above Ac_3 of the core and near, if not above, Acm of the case, as in (2), quenched, then reheated just above Ac_1, and again quenched. This first heating refines the core and dissolves most or all of the excess carbide in the case, but this first heating is overheating for the case and leaves it with too large grains. Reheating just above Ac_1, and again quenching, properly refines and hardens the case.

Treatment (4) gives the optimum mechanical properties, is applicable to steels of any grain size and to those with hypereutectoid cases. But the double quenching increases the likelihood of warping and cracking.

After any of these treatments the steel is given a low temperature draw to remove quenching strains.

The tendency is to use treatment (1) wherever feasible, as it takes the least time, fuel, and labor, and warpage troubles are slight.

Treatment (4) is the one most certain to give a non-spalling surface on a hypereutectoid case when this is produced so as to have excess carbide particles for wear resistance. Method (4) is sometimes modified for highly hypereutectoid steels by giving an initial quench instead of a slow cool from the carburizing temperature so as to avoid separation of a cementite network, which may redissolve with difficulty on the next reheat. This method of three quenches is too prone to produce warping to be widely applicable.

It will be noted that the most expensive treatment (4), cooling in the box followed by the double regenerative quench, is required by the steels that coarsen materially in carburizing, while the cheaper and less complex treatments are only applicable, with satisfaction, to the fine-grained steels.

Warpage. Every heating and quenching involves added tendency toward warping, which is almost always harmful on carburized parts, so one is often forced to use the single quench from the carburizing temperature. When only a single quench is contemplated, the optimum carburizing temperature is not over 1650°. The longer carburizing times required to get the proper depth of case at around 1600° lead to good diffusion and a smooth gradation from case to core.

Tempering after Quenching. Whatever the quenching cycle adopted, the steel after the final quenching is generally given a low-temperature heating for stress relief, but at so low a temperature that the case is not softened too much for the required service. Temperatures between 300° and 400° are used.

It is generally desired to leave the case file hard, which is normally accomplished by not exceeding 350° in tempering. The toughness of the case is increased by tempering. A condition often sought is that in which the case remains file hard but has had its Rockwell C or Scleroscope hardness dropped materially from that in the quenched condition.

Retained Austenite. The condition of file hardness plus relatively low Rockwell C or Scleroscope hardness may also be attained by the retention of austenite. As has been shown (Fig. 7, Chap. 6) austenite retention is favored by high C content, high alloy content, high quenching temperature, and by oil quenching compared with water quenching.

More austenite is retained in treatment (1) or (2), Fig. 1, than in (3) or (4).

Temperatures of Treatment. The temperature for treatment varies according to the criticals for the case and core; hence the Society for Automotive Engineers has listed suitable treatments for each of the steels on the S.A.E. list. The selection of alloy steels for carburizing will be discussed in Chap. 14, Vol. II.

The compositions for the various S.A.E. carburizing steels are:

	C	Mn	Ni	Cr	Mo	V	S
1010	0.05–0.10	0.30–0.60	....	....	....	....	0.055 max.
1015	.10– .20	.30– .60	....	....	....	....	.055 max.
X 1015	.10– .20	.70–1.00	....	....	....	....	.055 max.
1020	.15– .25	.30– .60	....	....	....	....	.055 max.
X 1020	.15– .25	.70–1.00	....	....	....	....	.055 max.
1115	.10– .20	.70 1.00	....	....	....	....	.075–0.15
1120	.15– .25	.60–0.90	....	....	....	....	.075– .15
X 1314	.10– .20	1.00–1.30	....	....	....	....	.075– .15
X 1315	.10– .20	1.30–1.10	....	....	....	....	.075– .15
2015	.10– .20	0.30–0.60	0.40–0.60	....	....	....	.05 max.
2115	.10– .20	.30– .60	1.25–1.75	....	....	....	.05 max.
2315	.10– .20	.30– .60	3.25–3.75	....	....	....	.05 max.
2320	.15– .20	.30– .60	3.25–3.75	....	....	....	.05 max.
2515	.10– .20	.30– .60	4.75–5.25	....	....	....	.05 max.
2512	max. .17	.30– .60	4.75–5.25	....	....	....	.05 max.
3115	.10– .20	.30– .60	1.00–1.50	0.45–0.75	....	....	.05 max.
3120	.15– .25	.30– .60	1.00–1.50	.45– .75	....	....	.05 max.
3215	.10– .20	.30– .60	1.50–2.00	.90–1.25	....	....	.05 max.
3220	.15– .25	.30– .60	1.50–2.00	.90–1.25	....	....	.05 max.
3312	max. .17	.30– .60	3.25–3.75	1.25–1.75	....	....	.05 max.
3915	.10– .20	.30– .60	2.75–3.25	0.60–0.95	....	....	.05 max.
4615	.10– .20	.40– .70	1.65–2.00	....	0.20–0.30	....	.05 max.
4620	.15– .25	.40– .70	1.65–2.00	....	.20– .30	....	.05 max.
4815	.10– .20	.40– .60	3.25–3.75	....	.20– .30	....	.05 max.
4820	.15– .25	.40– .60	3.25–3.75	....	.20– .30	....	.05 max.
5120	.15– .25	.30– .60	....	0.60–0.90	....	....	.05 max.
6115	.10– .20	.30– .60	....	.80–1.10	....	0.15 min.	.05 max.
6120	.15– .20	.30– .60	....	.80–1.10	....	.15 min.	.05 max.

(Silicon in basic open hearth steels 0.15–0.30. Electric and acid open hearth 0.15 min.)

In all cases the suggested carburizing temperature is 1650–1700°. The treatments are:

Steel	Carburize and Quench Direct	Carburize, Quench Direct, or Slow Cool and		Carburize, Slow Cool				Cyanide or Activated Bath *		Temper at
		Reheat to	Quench in	Reheat to	Quench in	Reheat to	Quench in	Heat at	Quench in	
1010 1015	Yes	1400–1450	Water	1650–1700	Oil or water	1400–1450	Water	1500–1650	Oil or water	250–325°
X1015 1020 X1020 1115 1120 X1314 X1315	Yes	1400–1450	Oil or water	....	....	....	Oil or water	1500–1650	Oil or water	250–325°
2015	Yes	1400–1450	Oil or water	....	....	....	Oil or water	1500–1650	Oil or water	250–300°
2115	Yes	1375–1425	Oil	1650–1700	Oil	1375–1425	Oil	1650–1700	Oil	250–300°
2315 2320	Yes	1350–1400 for case or 1450–1500 for core	Oil	1500–1550	Oil	1350–1400	Oil	1450–1650	Oil	250–300°
2515† 2512†	No	(Slow cool) 1325–1375 for case or 1400–1475 for core	Oil	1500–1550	Oil	1325–1375	Oil	....	....	250–400°

* Temper after cyanide or activated bath if desired.

† First normalize or anneal to desired structure.

Steel	Carburize and Quench Direct	Carburize, Quench Direct, or Slow Cool		Carburize at Same Temp., Slow Cool					Heat in Cyanide or Activated		Temper at
		Reheat to for Case	Reheat to for Core	Quench in	Reheat to	Quench in	Reheat to	Quench in	Bath at	Quench in	
3115* 3120*	Yes	1400–1450	1450–1500	Oil or water	1525–1575	Oil	1375–1425	Oil or water	1500–1650	Oil	250–300°
3215* 3220*	No	1375–1425	1425–1475	Oil	1525–1575	Oil	1375–1425	Oil	....	....	250–500°
3312†	Yes	1375–1425	1425–1475	Oil	1525–1575	Oil	1375–1425	Oil	....	....	250–300° 250–400° after direct quench
3415†	Yes	1375–1425	1450–1500	Oil	1500–1550	Oil	1375–1425	Oil	....	....	
4615† 4620†	Yes	1425–1475	1475–1525	Oil	1525–1575	Oil	1375–1425	Oil	....	....	
4815† 4820†	Yes	1350–1400	1450–1500	Oil	1550–1650	Oil	1350–1400	Oil	....	....	250–300° 250–500° after direct quench
5120† 6115† 6120†	No	1425–1475	1475–1525	....	1600–1650	Oil	1425–1475	Oil	1500–1650	Oil or water	250–350°

* First normalize above carburizing temperature for all save cyanide and activated baths.

† Normalize at least 50° above carburizing temperature. Anneal to desired machinability.

"Microstructures of SAE Steels after Recommended Heat Treatments," by Sauveur and Reed (published by E. F. Houghton & Company, Philadelphia), shows the case and core structure of the carburizing steels.

Austempering of Carburized Steels. It was brought out in Chap. 5 that austempering, quenching a steel from the austenitic field into lead or other fused bath at a suitable temperature, can be made to result in the formation of secondary troostite with a very interesting combination of hardness and toughness. This process is most applicable to high-C steels and to pieces of small cross section.

The case of a carburized steel is high C, and is thin. Hence it would appear that austempering might be utilized in the heat treatment of carburized steels, especially in case where maximum file hardness is not required for wear resistance. Because of the remarkable endurance properties of austempered steel (Fig. 14, Chap. 1), it is possible that producing an austempered skin on a machine part might greatly improve its endurance, especially since the highest stress on a part under repeated stress is usually at the surface.

Queneau and Dowdell[2] quenched several carburized steels into molten salt at 400°. Holding for 5 hr. gave a surface hardness of 60–62 Rockwell C. At 600° for 1 hr. 55 Rockwell C was obtained.

The steels used were one of 0.64 C, 0.80% Mn; 0.42 C, 0.77% Mn, 0.20% Mo; S.A.E. 2330 of 0.32 C, 0.65 Mn, 3.45% Ni, and S.A.E. 3120 of 0.18 C, 0.65 Mn, 1.20 Ni, 0.64% Cr.

It was necessary to use ½-in. square bars of the first two steels to get a case of 60 Rockwell C by austempering at 400° after box carburizing 8 hr. at 1700°, though 1-in. diam. bars of the latter two hardened at the surface almost the same as in oil quenching. Impact results on all the steels, carburized in the regular Charpy size, showed no real difference between quenching and tempering, and austempering. In the bend test on carburized ½-in. sq. bars the last two steels also behaved the same, irrespective of the treatment, but the 0.64% C steel, water quenched and tempered, supported about 1000 lb., while when austempered it supported 4000. The 0.42 C, 0.20% Mo steel, water quenched and tempered, supported 3000; austempered, 4000.

Queneau and Dowdell conclude that austempering carburized parts has possibilities as it allows the use of a steel of higher C content than could be used in the normal heat treatment of carburized steels, since, with the normal treatment, quenched samples would be weak because of the stresses set up in hardening.

These authors did not report on the properties of steels with cases austempered to 55 Rockwell C by quenching into salt at 600°. There are interesting possibilities in carburizing and austempering for strength, steel of either low or medium C content, especially when the case is brought to about 55 Rockwell C, that have not yet been explored.

Properties of the Carburized Piece as a Whole. When a very brittle case is attached to even a very tough and ductile core, and a crack starts in the case, it propagates through the core with ease, so that the whole piece tends to fail in brittle fashion under impact severe enough to crack the case. The crack would propagate with greater ease through a harder core; hence the choice of the very low-C steels for carburizing with heavy cases.

Even though the shock resistance of the piece is low under a single blow strong enough to break it, there is doubtless a useful and unmeasured degree of toughness in service of the carburized part having a core that is very tough when tested without the case. What is actually sought is resistance to notch propagation under repeated stress conditions that start a fatigue crack in the case. Single blow impact tests do not necessarily measure this property. A specific example of the need for core toughness is in carburized automobile tire chains, where the service involves abrasive wear and also repeated impact, from the slap of the chains. The core must not readily break by impact either before or after the case is worn off, and a tough core is best fitted for such service.[3]

The depth of the case comes into this notching effect, for, unless the case is thick enough to support itself against deformation, the soft core may fail to support it. This is analogous to a Brinell impression showing through on the under side of a thin specimen.

Support for Thin Cases. Obviously, another way to tackle the problem is to use a very thin case and to back it up by much harder material, and an alternative way of making gears is to use a higher C, fairly high alloy steel, and give it only the thinnest of cases, but have that case so thoroughly backed up that it does not crack. Such cases are usually applied by cyaniding, as will be discussed in Chap. 13, or by a brief gas carburizing treatment, discussed in Chap. 12.

Distortion of Gears. It is becoming accepted that most automotive gears are not subject to impact stresses,[4] that impact resistance of gears is far less important than fatigue resistance, particularly resistance to repeated overstress, and that the main trouble with automotive gears is the slight distortion that occurs on carburizing. Hence a wider range of carburizing steels is being considered for such use.

Conclusions as to the best carburizing steels for a given gear are greatly affected by the gear-cutting method already in use. By cut-and-try methods a contour is found, which, after carburizing and hardening, gives the least lapping-in to be done. Another steel and another treatment, with a different degree or type of distortion, will give very bad results, while the old contour is retained, but equally good results when the proper contour is found.

The conclusion that in these gears impact is a minor matter is not necessarily true for many other gears, or for many carburized parts in general.

Carburizing for Endurance. Any rotating shaft carburized or similarly strengthened on its surface will show increased endurance under repeated stress applied to the rotating beam, because the stress is at a maximum at the surface and is resisted by the stronger surface. In axial loading where the whole cross section is more nearly equally stressed, this improvement of endurance limit is nowhere nearly as marked.

In roller bearings, where the maximum compressive stress comes not at, but just below, the surface, fatigue failures may start within the core or at the junction of case and core of a carburized race or roller. Carburizing of antifriction bearings is, therefore, a special case in which depth of case and gradation of case to core, as well as the surface hardness and wear resistance, are of increased importance. In order to produce a harder type of carbide in the case, these steels normally contain Cr and sometimes also Mo, V, or higher Mn than usual, since all these are carbide formers.

The Human Element in Case Hardening. The factors of experience and personal judgment count heavily in all heat treatment, but nowhere more heavily than in the production of carburized parts. There are so many factors than can be intentionally or unintentionally altered which affect the final result that extreme care is required to ensure that the product is uniform.

Each different type of service has certain requirements as to hardness and toughness of the case and core, and as to the gradation between them. Such requirements are not easy for the engineer to state, nor is it easy for the metallurgist to measure the type of toughness involved. One may fairly question the real value of tests made on the core without the presence of the case, or of tests on bars carrying both case and core that are not of the shape used or subjected to the type of stresses that occur in service.

These difficulties are reflected in the varied opinions as to what steel should be used, what core hardness is needed to support the case, and what case thickness and carbon content should be secured. There is real unanimity only in that fine-grained steels afford superior service and that warpage is minimized by putting the steel through as few heating and quenching cycles as will serve. These facts combine to focus attention upon those fine-grained steels that are suitable for single-quench treatment.

After having selected a suitable steel and a suitable carburizing and heat-treatment practice, eternal vigilance is required to maintain the selected conditions. Examination of incoming stock, careful control of temperature, of the uniformity of the carburizing medium, and especially of the temperature uniformity of the charge are vital. Inclusion of pilot test pieces that are broken and examined for depth, and adhesion, of case and for hardness is essential as a final check.

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CHAPTER 12

CONTROLLED ATMOSPHERES

Scaling. The study of carburizing leads naturally to the consideration of controlled atmospheres in heat treatment. If steel is heated in a muffle to which air has access, or if, in an unmuffled furnace, the flue gases resulting from the economical combustion of fuel reach the hot steel, it oxidizes and scales.

Such flue gases from economical combustion have very little carbon monoxide (CO), much carbon dioxide (CO_2), and usually some free oxygen (O_2), that is, the fuel is burnt with a slight excess of air. The gases also contain a good deal of water vapor from the burning of the hydrogen (H_2) of the fuel.

The combustion engineer calls a flue gas containing excess O_2 "oxidizing"; one with no free O_2 and little or no CO, i.e., resulting from practically perfect combustion, "neutral"; and one with considerable CO, "reducing." But these terms are *not* correct in relation to the action of the gases on steel. Flue gas even from such smoky combustion that the fuel is wastefully burnt is definitely oxidizing to red-hot steel.

Smoky Flames. Scaling may be slowed down by use of smoky flames, but not eliminated. The water vapor and CO_2 in the products of even very incomplete combustion make flue gases oxidizing. Of course anything that reduces the amount of active oxygen (either free O_2 or that combined as CO_2 and H_2O), that reaches the steel, reduces scaling. Displacing the air or flue gases within an open furnace by putting in pieces of wood to char, charcoal to burn, dripping in oil vapor, or leading in a fuel gas which burns at the furnace vents will tend to lower the amount of active oxygen meeting the steel. Upthegrove[1] finds that regulating combustion to give 3% CO in flue gas will cut scaling, in direct firing, by 75%, but the 25% still remains. According to Jominy[2] it takes 15% CO in flue gas to bring about practical elimination of scale. Even this does not give "clean" hardening. A huge proportion of the fuel would be wasted in using so great a deficiency of air as to give 15% CO in the flue gas.

Smoke Burners. An obvious way to attack the problem is to lead the fuel gas into the furnace so that it surrounds the steel charge and does not meet the air for combustion till it passes beyond the charge. The gas will crack and deposit C on the work. This layer and the gas itself should protect the steel.

The incoming fuel gas is relatively cool when it hits the work; so the hottest gases of combustion do not wash the work, and much of the heating has to come by radiation from the furnace walls and from the hottest part of the flame itself. So-called diffusion burners have been devised to operate on such a principle. Obviously, any fluctuation in the gas currents within the furnace may displace the protective layer of gas and expose the work to scale-forming conditions. Fisher [3] appraises such burners as hard to control and prefers to separate the function of supplying the heating flame with gas from that of surrounding the work with raw gas. He advocates good premixing of gas and air for the heating burners and special "smoke" burners which carry a smaller amount of gas to the work, where that gas—unmixed with air—cracks and deposits soot. Such an arrangement he believes to be the most feasible for scale mitigation in unmuffled, fuel-heated furnaces.

Whether one should aim merely at the mitigation of scaling (taking care to avoid the production of a tight scale, difficult to pickle, and to avoid decarburization), and accept the necessity of pickling and grinding, or should seek complete elimination of both scaling and decarburization by elaborate controlled-atmosphere methods, is an economic problem to be decided by local conditions.

Mattocks [4] points out that where some scaling can be permitted, operating with about 0.50% excess O_2 over that required for complete combustion avoids decarburization.

The effort to avoid scaling, where the flue gas can meet the work, is usually attended with greater evil from decarburization than the scaling evil that is avoided or lessened.

In heating for hot rolling or forging, where the steel will later scale somewhat anyhow in the hot working (which of course is done in the open air), it will more often be the economical method to aim merely at mitigation rather than elimination of scaling in the heating process. In such service Conway [5] reports that efficient preheating of combustion air in an oil-fired, slab-heating furnace so that the time of heating is cut down, together with adjustment of flue gas to about 12% CO_2, makes the best compromise between fuel efficiency and scale thickness.

High-speed Steel. In the hardening of high-speed tool steel at 2300–2400° in the open, it is customary to gage the tool temperature by eye, noting when the scale starts to melt and blister. Besides the loss of metal, the irregular, blistered scale produces danger of irregular hardening in quenching. It is not easy to maintain muffles or metallic resistors at such a temperature; so controlled atmosphere furnaces for high-speed steel tend to be electrically heated by Globar resistors. The controlled atmospheres for such furnaces are generally produced by partly burning a fuel gas with a deficiency of air, controlling the ratio by manometers and valves, and feeding the mixture directly into the furnace without elimination of water vapor. The analyses of the gas atmosphere given in the literature tell only the free O_2, the CO_2, and CO, disregarding the important factor of the water vapor, which will, of course, vary greatly with the H_2 content of the particular fuel gas used.

Recommendations for gas composition on the basis of CO_2, CO, and O_2 usually agree in advocating complete freedom from unburned O_2 for the high-speed hardening heat, but vary considerably as to the proper ratio of CO to CO_2. Conflicting statements as to effect of the atmosphere upon the rate of heating and the propensity to grain growth in various atmospheres are also made so that the whole matter should be handled by experimentation on the actual high-speed steel and the actual fuel gas to be used. Broadly speaking, adjustment of the atmosphere to be free from unburned O_2, i.e., so that wood will char but not burn in the furnace, is helpful in minimizing scale. If water vapor, which tends to decarburize, is present, then a thin scale is probably desirable to protect the surface from decarburization and such a scale is produced when the atmosphere contains some CO_2.

There would seem to be little objection to giving a high-speed tool a thin skin slightly higher in C, but in practice this is seldom attempted save by so-called "pack-hardening," i.e., heating in charcoal. This is often applied, for example, to high-C, high-Cr tool and die steels, more rarely to high-speed. When pack hardening is applied, the steel is heated to only about 2000° to avoid excessive carburization.

The usual effort for high-speed steel is to get a controlled atmosphere that contains no free O_2 and has 6 to 12% CO, with corresponding amounts of CO_2, i.e., an atmosphere that results from slightly incomplete combustion and is definitely oxidizing at this temperature. The moisture is not ordinarily removed from the gas in practice. However, Peck [6] remarks that, in heating high-speed steel for quenching, *both* CO_2 and H_2O *must* be absent if decarburization is to be entirely prevented.

The high-speed steel preheat at 1400–1600° is generally carried out either in a definitely oxidizing atmosphere in order to form a thin adherent scale that will keep the high-heat atmosphere away from the metal itself, or in cast-iron borings. Because full muffle and electric furnaces are seldom used on high-speed steel, adjustment of the flue gas of the direct flame method is all that is ordinarily attempted in the high heat. When muffles and electric furnaces come into greater use for high-speed steel, slightly carburizing atmospheres will probably be found applicable, provided they do not injure the muffles or resistors.

Gill[7] finds that in heating high speed for forging, it is better to scale than to run the risk of carburizing or decarburizing, though were an atmosphere available whose action would be truly neutral for the time required in the slow heating and long holding necessary it would of course be better. For annealing, a lime or sand pack with only a small amount of charcoal is suggested. In heating for hardening, Gill states, the preferred controlled atmospheres produce a surface carburization, the C content rising some 0.25% C for a depth of about 0.001 in., which is normally removed in finishing the tool. Were the conditions decarburizing a much thicker soft skin would have to be removed.

Other Tool Steels. At the lower temperatures of C and low-alloy tool steel hardening there is some effort to do really bright hardening by using a definitely carburizing atmosphere, such as that from cracked oil vapor. In more common practice a mildly but definitely oxidizing atmosphere is chosen, to minimize scaling while avoiding decarburization. Palmer[8] advocates an atmosphere with some 10% free O_2 for various high-C, low-alloy tool steels, only selecting atmospheres with no free O_2 and with 12% or more CO for high-C, high-Cr and for Cr-W, hot-work steel. Such recommendations are not made to give the greatest avoidance of scale but rather with the aim of avoiding decarburization while accepting considerable scale formation.

Palmer[8] finds some interesting things about the effect of atmosphere on the toughness of C tool steel (1.10% C), with toughness evaluated by a torsion impact tester. The atmospheres studied are not stated in detail but are referred to as "oxidizing, neutral, and reducing" in the phraseology of the combustion engineer, which probably means that they are all oxidizing in varying degree. There is not a huge difference between the torsion impact results on a given type of steel heated in these various atmospheres, and the fine-grained "tough timbre steel" stands heating in any of the atmospheres studied by Palmer with very little loss of toughness compared to the result obtained in a lead bath. The coarse-grained "brittle timbre" steel, however, is strongly affected by all the atmospheres, while if heated in the lead bath it is quite as tough as the "tough" steel at normal hardening temperature, and only slightly worse at still higher temperatures. (d'Arcambal remarks[9] that makers of small tools ordinarily heat in lead baths.)

Thus the coarse-grained steel is shown to be seriously affected by surface attack. This peculiar phenomenon is not well understood nor satisfactorily explained. An analogous phenomenon, by which it has been alleged that, in the proper atmosphere, high-speed steel will stand exposure times in heating for hardening without grain growth that would, in the ordinary, more oxidizing atmosphere, produce ruinous coarseness, has received much discussion. The mechanism involved is explained in several ways, none of them entirely convincing. However, these observations indicate that subsidiary effects may appear, arising from what is done to the surface by atmospheric attack during heating, so that atmosphere control may have more far-reaching results than appear at first sight.

Heating Ordinary Steels for Hardening. Heating in a muffle or electric furnace that is kept closed so there is no draft through it tends to restrict the oxygen meeting the steel to that originally in the furnace when the steel was charged. Such furnace atmospheres are not really controlled; they are merely mitigated in order to reduce scaling somewhat. The surfaces produced require machining or pickling to remove the scale that does form.

The scaling of steel is greatly increased by the presence of SO_2, from fuels containing S.

Scale *vs.* Decarburization. Precautions to cut down scaling, without entirely preventing it, often run into two difficulties—the thinner scale may be tighter and more difficult to remove on pickling and, when too thin a scale is made, there may be a decarburized layer beneath it. A soft, decarburized bark is to be avoided in cutting tools, where to have the cutting edge, or in springs, where to have the most highly stressed surface soft and weak, would ruin the performance of the piece.

For example, Boegehold[10] points out that in "knee action" springs for automobiles the steel, e.g., S.A.E. 9260, must be heat treated to 430–480 Brinell in order to stand the maximum applied stress of more than 115,000 lb./in.^{2} in torsion. If the surface is decarburized, its strength is only some 30,000 lb./in.^{2} in torsion; so surface cracks form and are propagated through the stronger core. Hence the steel, after hardening, must be ground to remove the decarburized bark, or else decarburization must be prevented by using a mildly carburizing atmosphere during heating for hardening. Similar arguments apply to axle shafts.

When scaling is heavy, oxidation progresses so fast that Fe is removed as fast as C. At a slower rate of scaling, C may be removed faster than Fe and faster than diffusion of C from the inside can take place to replenish the depletion of C at the surface. Water vapor tends to decarburize rapidly, as it breaks down at the surface of the steel, the O combining chiefly with Fe, the hydrogen being freed to attack the C, forming CH_4. Jominy[2] cites an interesting case where decarburization was met in an electric furnace, heating gears for hardening, when a breeze from an open window blew steam from a quenching tank into the furnace. In controlled atmospheres free O_2 and SO_2 are the most fatal, H_2O the next most dangerous constituent.

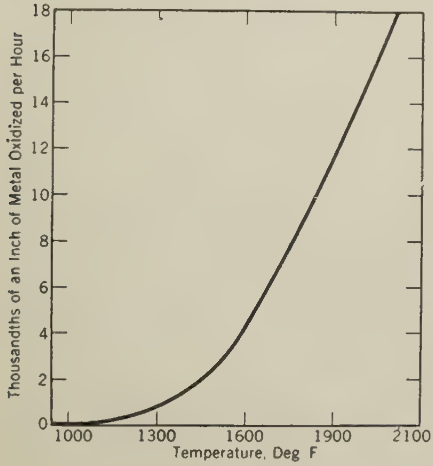


FIG. 1.—Rate of scaling in air, with increase in temperature. (After ref.[11])

Effect of Time and Temperature on Scaling. The rate of scale formation increases greatly with temperature, and of course scaling increases with time at a given temperature. The curve for ordinary steels is of the type shown in Fig. 1.[11] Tempering after hardening is usually done at such low temperatures that the resultant temper color or incipient scale is too thin to do harm, and tempering furnaces often circulate air over the work by a fan to secure better temperature uniformity.

On account of the time factor, scaling can be minimized by heating rapidly, but, as will be brought out in Chap. 14, rapid heating is undesirable since the temperature of the inside and outside should be as closely as possible the same before forging or quenching. Only in the case of those types of electrical heating where the heat is generated in the metal itself can rapid heating be accomplished and temperature uniformity be obtained. This internal heating of steel by its own resistance is feasible only in regular sections. Rods, tubes, and the like can be heated by passing heavy currents through them, as in the Snead process and in Berwick heaters, and the heating done so rapidly that even in air there is very little scale formation. High-frequency induction heating as applied to circular sections, like the bearing surfaces of crankshafts and piston pins, can be used to heat the surface rapidly and uniformly, without such a degree of scaling that would be harmful. Indeed, the crankshaft hardening process produces little more than a slight tarnish. Moderately irregular sections, like cam shafts, may also be hardened by induction heating.

Blueing. An oxidized surface may not always be detrimental. Production of a definite thin, adherent scale, followed by H_2 reduction just before hot galvanizing, is utilized in one process[12] to form a clean matte surface to which the Zn coating will adhere strongly. Intentional blueing, somewhat analogous to the old Bower-Barff coating,[13] produced by low-temperature heating in an atmosphere made oxidizing, as by admission of steam, is produced by admitting steam into an annealing box during the cooling cycle and is utilized in producing a commercial finish.[14] Safety razor blades are produced with a blue finish. Blueing may be produced by air at low temperatures, as a "temper color," or in a salt bath compounded to have a controlled oxidizing power. A tenacious oxide film is sometimes produced on cast-iron piston rings, to cut down wear.

Heating to 1200° for an hour or so, in spent carburizer or in charcoal, followed by oil quenching, produces a thin, adherent scale, indicating that the $C:CO:CO_2$ equilibrium for this temperature makes the gas atmosphere slightly on the oxidizing side for steel.

Annealing. At the annealing temperatures used to soften work-hardened, low-C steel, e.g., 1200–1250°, scaling may be of minor importance on heavy stock, but it is important on very thin sheet. Cold-rolled sheet or strip is annealed in packs or in coils; and, in order to drive the heat through the air spaces between the leaves of metal, since such air spaces are good heat insulators, very long heating periods must be used, and almost equally long cooling periods are required before the stock can be handled. A cycle may take 40 hr. Although the temperature is not high, if air is present the edges of the sheet will be badly scaled. When a clean surface is required, the pack or coil is subjected to "box" annealing, the stock being enclosed in a covered container heated from the outside. The "box" or "bell" is usually placed over the pack or coil (which may weigh several tons) and sets into a sand seal. On heating, the air expands and some of it escapes, the remainder attacking the steel. To prevent this, some non-oxidizing gas may be led into the box to displace the air. It is especially important that the gas be allowed to flow through it during the cooling period to prevent air inflow as the enclosed gas contracts on cooling.

Introduction of city or natural gas in box annealing was the first use of a "controlled atmosphere" in heat treatment. Raw fuel gas will deposit C on the work as it cracks, so the next step was to crack it outside and lead in the H_2-CH_4 mixture so produced. Still later, other protective gases, such as H_2, or cracked NH_3, or atmospheres specially prepared by burning a fuel gas with a deficiency of air, came into use. The H_2 and the 75% H_2, 25% N_2 atmospheres (the latter produced by cracking anhydrous ammonia, NH_3) find some use commercially in annealing of silicon transformer sheet.

Bright Normalizing. The majority of "controlled atmosphere" applications in use are in the box or bell annealing of very low-C sheet and strip (see Figs. 2 and 3), and the atmospheres are produced from some readily available fuel gas and air. Since the advent of continu-

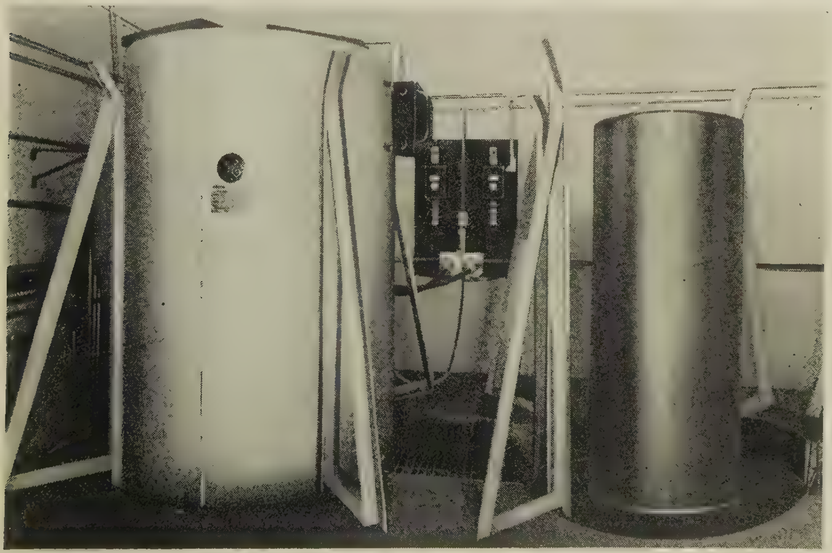


Fig. 2.—Electric bell annealer for coiled wire. The coils are piled and covered with the gas-tight container shown at the right. The protective gas atmosphere is led into this container. The furnace itself is at the left. The heating elements are on the inner wall. The furnace is lifted off the first container, by a crane, and set over a second container so that the second may be heating while the first is cooling. (Courtesy General Electric Co.)

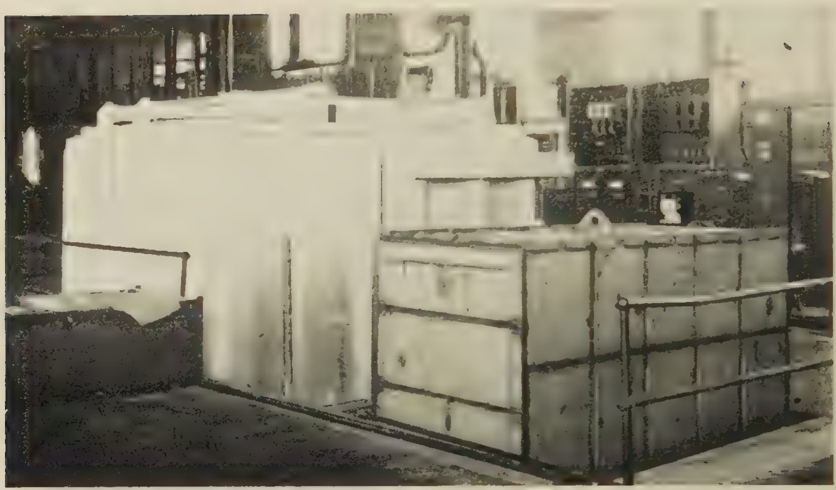


FIG. 3a.—Electric box annealer. (Courtesy Westinghouse El. & Mfg. Co.)

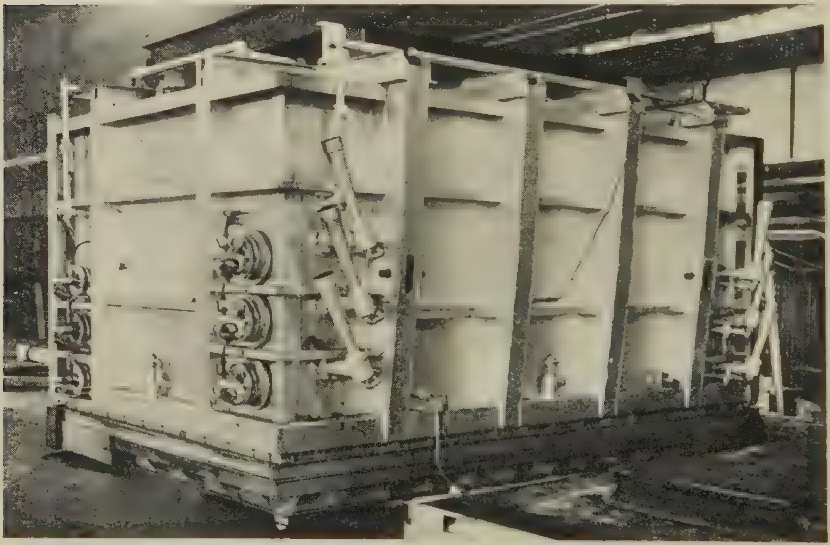


FIG. 3b.—Radiant tube box annealer. (Courtesy Surface Combustion Co.)

FIG. 3.—Box annealers. Stacks of steel sheet are covered by a rectangular box, within which the protective atmosphere is led. The furnace, with resistor elements, as in (a), or with radiant tubes, as in (b), is lifted off or on the inner cover for cooling or heating.

ous strip and wide strip mills, continuous normalizing of cold-rolled, low-C steel in controlled atmospheres has also become important. Here the temperature may go up to 1750° but the time at temperature is short, of the order of 10 to 15 min. Because of the great speed at which the strip travels, the furnaces may be very large. One at the Ford Motor Company is 330 ft. long, including the primary heating, cooling, reheating, and final cooling zones. (Compare Fig. 4.) Both the bright annealing and the bright normalizing processes normally relate to very low-C, plain C steels.

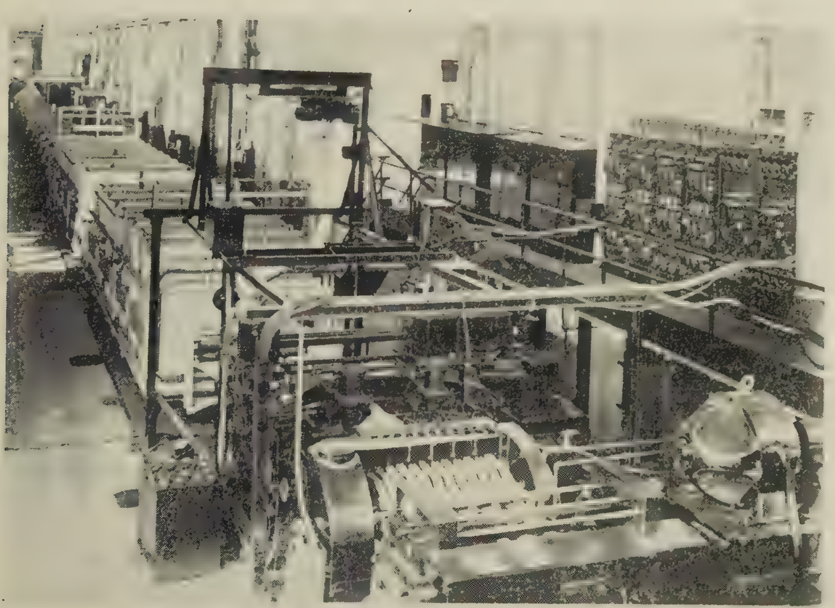


FIG. 4.—Continuous, controlled-atmosphere, electric normalizing furnace at the Ford Motor Co. This furnace is 230 ft. long. Another similar furnace in the same plant is 330 ft. long. (Courtesy Westinghouse El. & Mfg. Co.)

Spheroidizing annealing of high-C steel wire is also being done in controlled atmospheres, but *not* in the same atmospheres that are used for low-C steels.

Clean Hardening. The success of bright annealing and bright normalizing, with avoidance of the messy pickling process and the loss of steel involved, has made the heat treater anxious to do "clean hardening" on his medium- and high-C plain and alloy steels. Hardening furnaces provided with means for surrounding the work with a suitably controlled atmosphere are on the market or are specially designed in the case of large continuous furnaces. The atmospheres employed are still far from standardized, nor are they wholly satisfactory from all points of view, including performance, ease of control, and cost. It is easy enough to avoid scaling, but to do so without decarburization is far less easy.

Furnaces for Controlled Atmospheres. Any controlled atmosphere furnace must employ the "diving bell" principle of a container sealed against access of air or flue gases, with inlet and outlet to carry a stream of the desired gas at slightly above atmospheric pressure to avoid infiltration of air. This is not difficult in a batch furnace. To allow charging and discharging of a continuous furnace without inward leakage of air, special seals, a more rapid gas flow, and a higher pressure are required. (See Figs. 5, 6, and 7.) The container must be externally heated, like a muffle furnace, or, if internally heated, the heating must be by electric resistors or radiant tubes so that no products of combustion reach the inside of the container. Care must be taken to flush the air out of the furnace after a shut-down, before charging the steel.

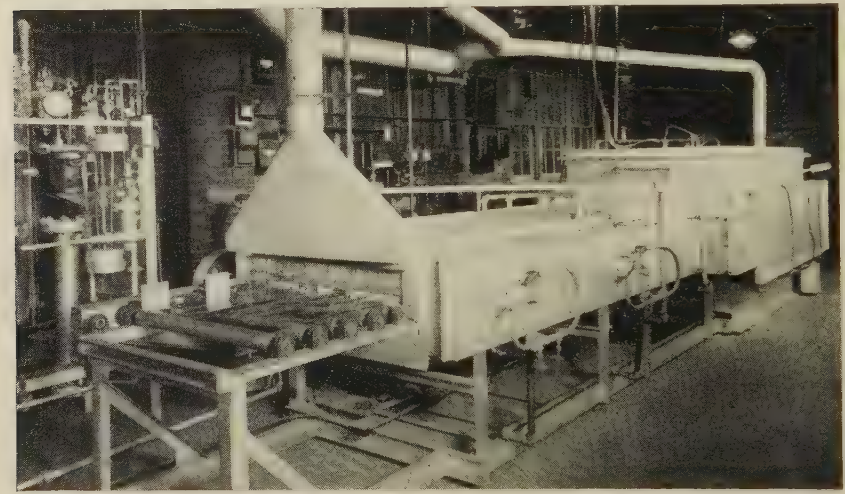


FIG. 5.—Continuous, controlled-atmosphere, normalizing furnace, operating on cold-rolled strip. (Courtesy Westinghouse El. & Mfg. Co.)

Gases Used or Proposed. Given such a furnace, the problem resolves itself into providing a gas that will not scale or decarburize a given steel at the temperature and time required for heat treatment.

Nitrogen. Pure dry nitrogen (N_2) would be relatively inert to the work under most conditions (though at unusually high temperatures even a plain C steel can pick up some N_2). Some resistor alloys and some heat-resisting alloys, unless their surfaces are completely protected by a preformed oxide coating, are quite readily nitrided in, and hence deteriorated by, N_2. Besides its cost, there is the difficulty of getting commercial N_2 entirely free from traces of O_2 or CO_2, which are undesirable, nor does a wholly neutral gas provide any mitigation of the effect of infiltered air.

Jominy [15] remarks that dry N_2 with as little as 0.1% O_2 will decarburize worse than air. Bramley and Allen [16] (see also Jominy [2] and Webber [17]) have shown that N_2 which contains the 2% of water vapor that it can hold at 65° is badly decarburizing while even the 0.05% it can hold at —20° has a distinct action. (Compare Fig. 8.)

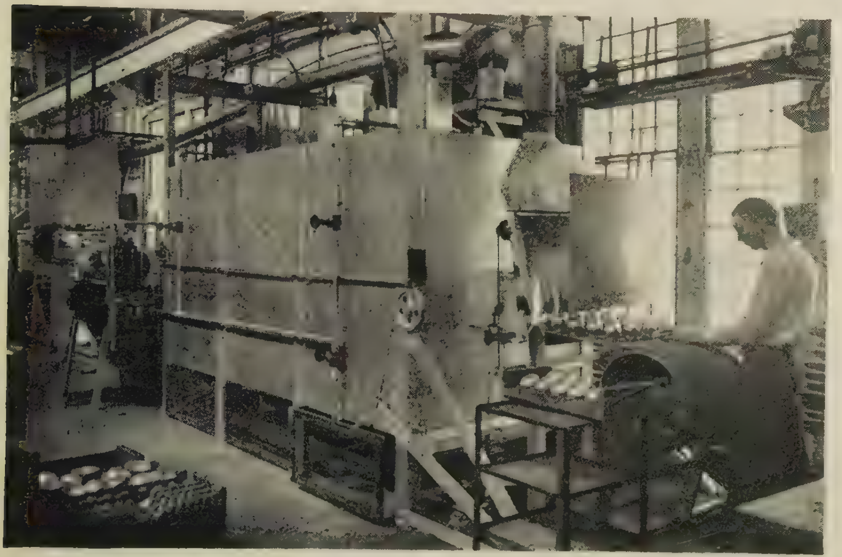


FIG. 6.—Continuous, controlled-atmosphere, brazing of balls at a plant of the General Electric Co. The furnace may also be used for bright annealing.

Proposals to develop commercial N_2 generators to provide a controlled atmosphere, aiming to be "pure" N_2, should be very carefully examined. An "impure" N_2 carrying small amounts of some gas, such as CO, H_2, or CH_4, that will oppose the decarburizing effect of traces of H_2O, would appear much to be preferred.

Most commercial gases suggested for controlled atmospheres do contain appreciable amounts of the "reducing" gases H_2 or CO, or both, and perhaps of CH_4. These reducing gases are likely to be accompanied by the oxidizing gases H_2O and CO_2, so that a balance has to be sought. Unfortunately, this balance is a variable one, depending on the composition of the steel to be handled, as well as upon the temperature at which heating is to be carried out.

Effect of Controlled Atmospheres on Carbon and Alloy Steels. The main factor governing the behavior of a given gas as to staining, and to carburization or decarburization, is the C content of the steel. Broadly speaking, the moderately alloyed steels, of the S.A.E. series, for example, act very much like a plain C steel of the same C content. This is not exactly true, but it serves as a good general approximation, as the work of Slowter and Gonser [18] shows, in respect to carburization and decarburization. In respect to staining, there is one notable exception to this general rule. Steels high in Cr are stained by atmospheres containing CO, which do not stain C or other alloy steels. The low and medium Cr S.A.E. steels for heat treatment have a tendency to

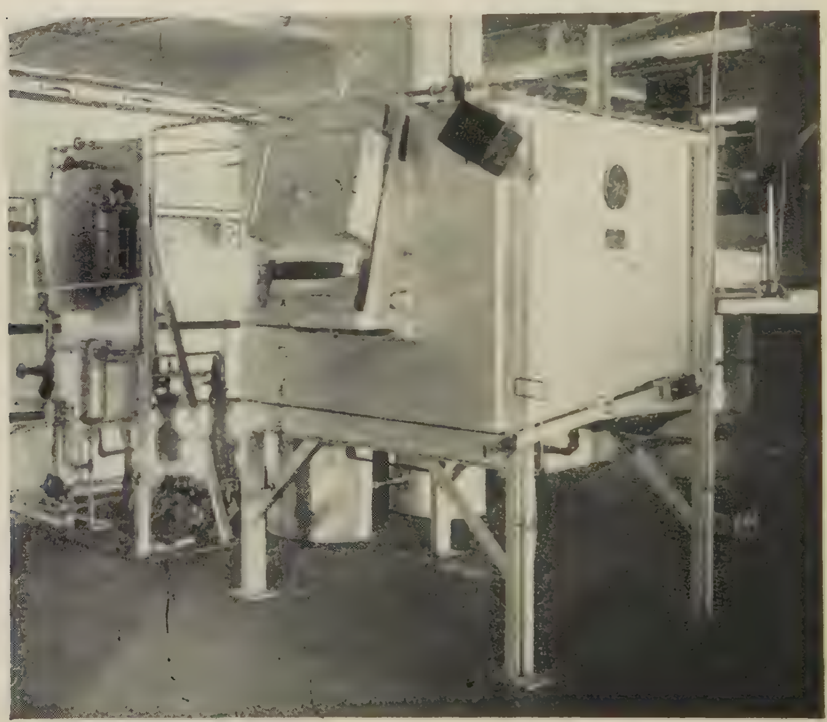


Fig. 7.—Batch type controlled-atmosphere furnace. The gas generator in which fuel gas and air are burned to produce the controlled atmosphere, together with equipment for condensing and removing most of the moisture, is shown at the left. (Courtesy General Electric Co.)

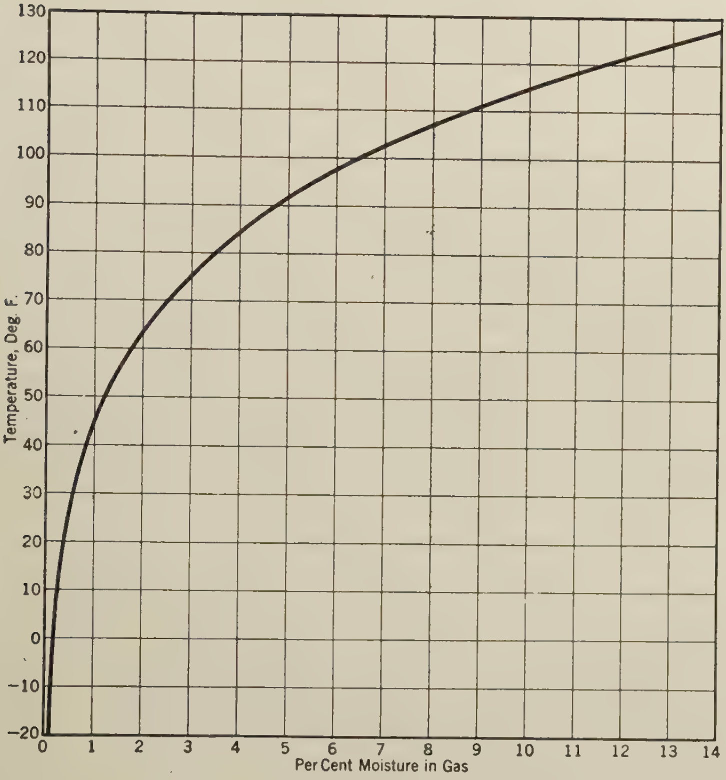


FIG. 8.—Dew-point Curve. A wet gas, cooled to the temperatures shown, will deposit its excess moisture, but will retain the amount shown. A wet gas can be dried by a water spray whose temperature is below that of the gas. Its final moisture content will be that of the temperature to which it is cooled by the water. If the cooling water, brine, or other medium does not contact the gas but is carried in pipes, the moisture condensed from the gas will collect on the pipes. To allow it to drip off, instead of forming frost, the temperature must be above 32°. Cooling coils are therefore used only down to about 35°, corresponding to 0.6% H_2O. For small H_2O content, the gas may be passed over activated alumina, by which the H_2O content can be brought down to about 0.01%.

be stained in atmospheres containing CO, but with such amounts of Cr the oxide film is something like a temper color on a C steel and is so thin and uniform that it will probably exert little if any effect on the rate of cooling in quenching.

Were it necessary to bright anneal the ordinary S.A.E. steels containing Cr, the atmospheres containing CO would need to be avoided along with those containing free O_2, CO_2, or H_2O, but this avoidance of CO is not required in hardening. The high-Cr steels like 18 Cr 8% Ni stainless cannot be kept bright when heated at the usual temperature in gas containing CO. For this purpose only completely dried H_2 or $H_2 : N_2$ will serve, and the merest traces of O_2, H_2O, CO_2, or CO must be avoided.

From the point of view of heating ordinary alloy steels for quenching in order to avoid decarburization, one may consider that the behavior will be that of an ordinary C steel of the same C content.

Decarburization. Unfortunately H_2 and CO are not merely *reducing* gases. Hydrogen always and CO under certain conditions are *decarburizing* gases, so that a gas mixture containing H_2 or CO that is just balanced to prevent oxidation is decarburizing, and a further balance must be made to get out of the decarburizing field and onto the neutral line where decarburization may be avoided.

Before considering the more complex gas mixtures, the two simplest cases, and the ones most directly related to gas carburizing, should be discussed. The active agent in several of the gases used for ordinary gas carburizing is methane, CH_4, either supplied as such in natural gas or produced by decomposition of higher hydrocarbons into $CH_4 + H_2$. In ordinary box carburizing the active agent is CO, accompanied by a little CO_2

Hydrogen-methane Mixtures. If the $CH_4 : H_2$ mixture is very low in CH_4, it will not carburize steel. Pure H_2, even if dry, has a decarburizing action on any steel; the decarburizing action is stronger the higher the steel is in C. That is, there is a reversible reaction,

$$Fe_3C + 2H_2 \rightleftarrows 3Fe + CH_4$$

whose tendency to go in one direction or the other is affected by temperature and by the amount of C already present in the steel. Thus, for any temperature in the heat-treating range, it is possible to find, for steel of any C content, a $H_2 : CH_4$ ratio which will be neutral to the steel, i.e., will neither carburize it nor decarburize it. Other hydrocarbons such as propane, butane, etc., can be used, with certain limitations as to deposition of soot. Since these tend to crack into soot, CH_4, and H_2, CH_4 may be considered as representative of hydrocarbons in general.

The relation can be expressed in the plot of Fig. 9, from data by Sykes.[19] This shows, for example, that at 1650° a gas containing only one part of CH_4 to 50 parts H_2 will be neutral to a 0.95% C steel, will carburize steel of lower C content, and decarburize those of still higher C. Naturally, when the $CH_4 : H_2$ ratio and the temperature are both fairly close to the equilibrium curve of Fig. 9 for a steel of a given C content, the action, either carburizing or decarburizing, is slower than when the ratio and temperature are far away from the curve for equilibrium. Dilution with an inert gas such as N_2 also slows down the action but does not alter the final equilibrium. Commercially, instead of pure H_2, cracked anhydrous NH_3 is more commonly used because of cost. Cracked NH_3 contains 75 H_2, 25% N_2. Since it is the $CH_4 : H_2$ ratio that counts, in adding CH_4 to cracked NH_3 to combat decarburization one needs to add only three-fourths as much CH_4 as would be required with pure H_2.

In the case of H_2 and CH_4, with or without N_2 (in the absence of contaminating gases that contain active O), there is no problem of scaling or oxidation. Only one action, that of carburization or decarburization, needs to be considered. (This is not true if the gas contains H_2O, and we shall discuss this complication later. Soot deposition also has to be taken into account.) But no $CH_2 : H_2$ ratio is inert to all steels at a given temperature, or to any one steel at different temperatures. For the gas to be inert to a given steel at a given temperature, the right $CH_4 : H_2$ ratio must be used. Figure 9 gives information necessary for choice of the correct ratio.

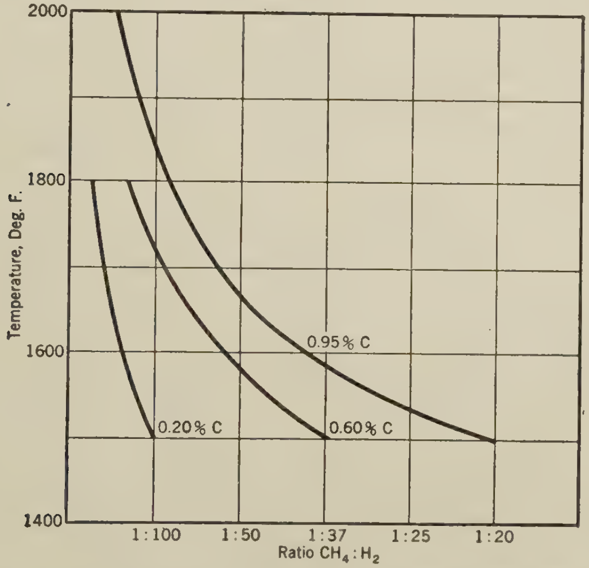


FIG. 9.—Effect of various ratios of CH_4 to H_2 on steels of various C contents at various temperatures. A composition to the right of a curve is carburizing to that steel; a composition to the left is decarburizing. (After Sykes.[19])

Carbon Monoxide-Carbon Dioxide Mixtures. The simple gas pair $CO : CO_2$ (with or without some dilution by N_2), when a reactive form of free C is present, and especially when an energizer like BaO is also present, is carburizing to steels of all commercial C contents, at ordinary carburizing temperatures, as has been discussed in Chap. 10. This is because the reaction $CO_2 + C \rightleftarrows 2\,CO$ at these temperatures, and especially with the energizer present, produces a gas mixture high in CO and low in CO_2. If the free C is left out, or if a non-reactive form of C is used, and the energizer omitted, the CO is scarcely high enough to carburize a eutectoid steel, and such conditions are used in box carburizing when the maximum C content in the case is to be held down.

If the CO_2 of the CO : CO_2 mixture is too high, such a ratio will decarburize instead of carburize. Just as in the case of $CH_4 : H_2$, the CO : CO_2 ratio must be the correct one for the C content of the steel and for the temperature in question, and analogous equilibrium curves may be drawn. Such curves were shown in Fig. 13, Chap. 10, in the discussion of box carburizing. Figure 10 shows them in different form, as collected by several authors in the A.S.M. Carburizing Symposium. Putting the information in tabular form:

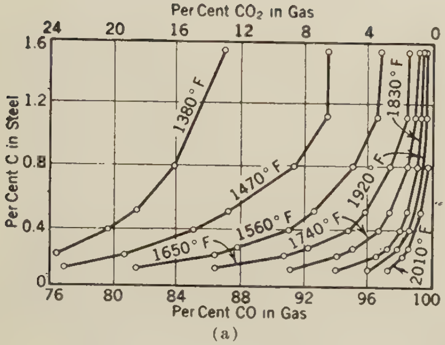
PERCENT CO_2 (BALANCE IS CO) AT EQUILIBRIUM IN CONTACT WITH STEEL

(Becker[21])

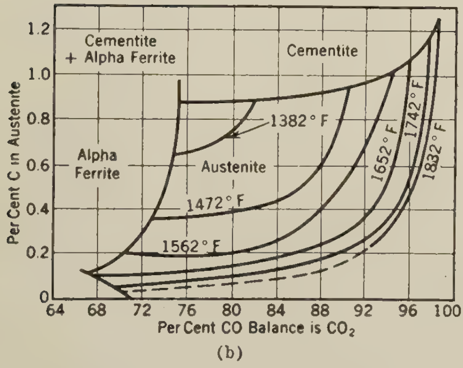
Temperature Deg. F.	Graphite	1.2% C Steel	0.8% C Steel	0.4% C Steel	0.1% C Steel
1200	62.6	55.7	55.7	55.7	55.7
1380	26.3	21.8	25.1	28.5	28.5
1560	5.85	5.7	7.1	12.5	26.7
1740	1.5	2.0	2.9	4.8	14.2

If one were to attempt to box carburize with graphite, one would find that when the amount of CO present is greater and that of CO_2 less in contact with graphite than with a given steel, the steel will be carburized. When the opposite occurs, or the CO_2 in equilibrium with graphite is greater than it is with a certain steel, decarburization occurs. By use of a more reactive form of C and of an energizer, the CO ratio can be slightly increased; but the lesser amount of CO_2 in equilibrium even with a reactive carburizing compound than with steel, accounts for the fact, noted in Chap. 10, that at 1200° carburized steel will be decarburized under the conditions obtaining in the box. Hence the desirability of rather rapid cooling of the box in carburizing, so as not to allow time for appreciable decarburization.

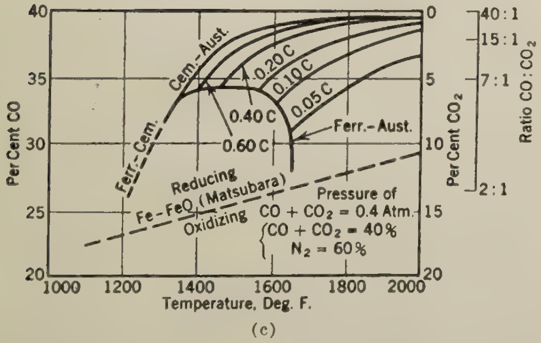
Since the commercial CO : CO_2 mixtures are diluted with N_2, one



(a) CO : CO_2 equilibrium with steel. (After Bramley and Lord, quoted by Grossmann.[20]) Gas of a ratio to the right of a curve is carburizing, that of a ratio to the left is decarburizing, for the temperature and C content of steel shown.



(b) CO : CO_2 ratios equilibrium with steel, as shown by Takahashi, quoted by Webber.[17]



(c) Equilibrium CO : CO_2 ratios in gas mixtures containing 60% N_2. After Johansson and von Seth, quoted by Webber.[17]

FIG. 10.—Various methods of plotting the relationships among gases of different CO : CO_2 ratios with steels of different C contents. See also Fig. 13, Chap. 10.

must remember that it is the ratio of these gases that counts, i.e., an 18% CO, 2% CO_2, 80% N_2 mixture, and one of 90% CO, 10% CO_2 will have practically the same action. Dilution with N_2 will slow down the action but will not markedly alter the final equilibrium. Hence, what appears to be low CO_2 in percentage may be high CO_2 as a ratio.

Scaling by Carbon Dioxide. If the CO_2 ratio becomes large enough, another reaction sets in; the gas becomes oxidizing and will scale steel heated in it. This behavior is shown in curve (C) of Fig. 11, in which the C content of the steel is not shown, the curves being for steel saturated with C at the various temperatures. With a $CO_2 : CO$ ratio on the right of the line *C*, i.e., with CO_2 predominating, steel can be scaled without the presence of any other gas than CO_2 and CO. At the left of *C* the gas no longer scales, and it reduces FeO, but it reacts with Fe_3C and decarburizes. Not till the CO is so high that we reach the line *D* does the gas become neutral of the steel. To the left of *D* the gas becomes carburizing.

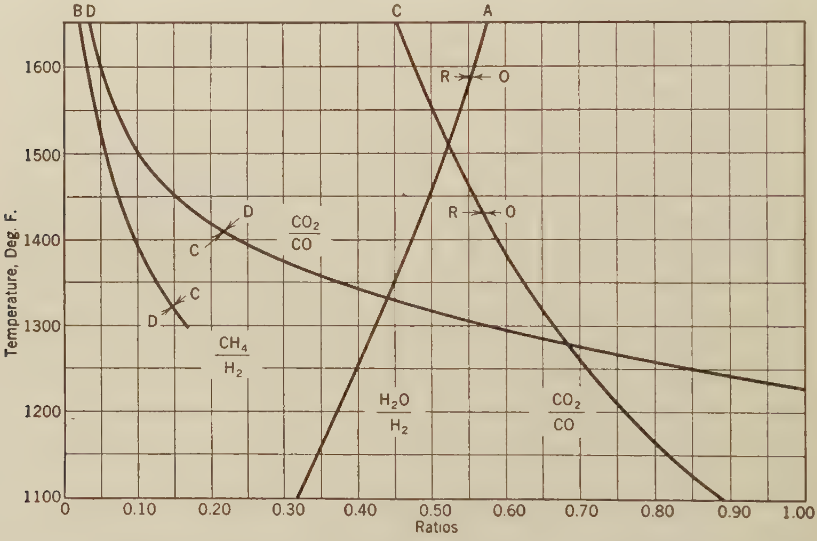


Fig. 11.—Equilibrium ratios for the $CH_4 : H_2$, : CO, and $H_2O : H_2$ reactions in contact with Fe-C alloys saturated with C at the temperature in question. (Stansel.[22])

O = Oxidizing
R = Reducing

C = Carburizing
D = Decarburizing

Inactivity of Pure Carbon Monoxide. From the point of view of carburizing, it should be remarked that, though these equilibrium curves *C* and *D* of Fig. 11 for $CO : CO_2$ indicate that pure CO or pure

CO plus N_2 would be strongly carburizing, actually CO does not readily break down to deposit soot and form CO_2 by the reaction $2\ CO \rightleftarrows C + CO_2$ and it appears essential, in order to have an actively carburizing gas, that a reversible chemical reaction should be going on, to produce nascent C. Thus at 1700-1750° a pure CO or pure $CO + N_2$ gas does *not* show a rapid rate of carburization.

The behavior of any $CO : CO_2$ mixture, diluted only with N_2 and free from any other active gases, can be predicted from Figs. 10 and 11, for steel of any C content at any heat treating temperature; and, were the commercial gases pure, controlling such an atmosphere would be simple. Figures 10 and 11 relate to stagnant gases. Varying rates of flow, especially at exposure times too short to reach true equilibrium, will somewhat modify the results, but the general relations are quite accurately shown by the curves.

Hydrogen-Water Vapor. Commercial gases are not pure; moisture, i.e., water vapor is present unless the gas has been specially dried to remove it. We may first trace the behavior of H_2O in presence of H_2. Such a mixture is *always* decarburizing. Even dry H_2 is somewhat decarburizing, and the reaction $Fe_3C + 2\ H_2 \rightleftarrows Fe + CH_4$ is accelerated and forced to the right by presence of H_2O.

In respect to scaling, H_2 may carry considerable H_2O at high temperatures and still be non-oxidizing; but, as the temperature drops, the tolerance for H_2O decreases. This is shown by curve *A* in Fig. 11, after Stansel,[22] and, at still lower temperatures, in Fig. 12, after Schenck.[23] Thus behavior is of commercial importance in bright annealing, when the steel has to be cooled down in the protective gas till it can be taken out in the air without danger of oxidation. In cooling steel that has been annealed or normalized in an atmosphere of H_2, a $H_2O : H_2$ ratio of 1 : 200, as Fig 12 shows, will produce staining at 400° if sufficient time is given; and, from experience with temper colors, we know that steel must be cooled to about this temperature before it can be exposed to the air without oxidation.

Hence, unless the steel can be quenched while still in the protective gas, moisture has to be carefully avoided during the cooling period, if all traces of staining are to be avoided. Thum [24] takes 1 : 20 as the limiting ratio in practice for bright annealing, i.e., in a gas with 10% H_2 not more than 0.5% H_2O can be tolerated. To reach that H_2O content by cooling a moist gas, it must be cooled below freezing, as is shown in Fig. 8; i.e., drying by activated alumina would be required rather than spray cooling.

Decarburization by Simple Gases. Experimental evidence on the behavior of some of the simple gases in relation to decarburization on heating for hardening is available from the work of Slowter and Gonser.[25] In this work the gas in question was passed over a series of steels ranging in C content from about 0.05 to 1.10%, usually for a period of 3 hr. at 1740°, and the steels were allowed to cool in the gas. The lower C steels were located at the gas inlet. The gas composition may be affected by the presence of a steel of a C content with which the gas is not in equilibrium; but, since the steel surface exposed was small and the gas was passed through in a steady stream, the results, although somewhat modified by the presence of steels of more than one C content, will serve as a first approximation. The temperature used was higher and the time much longer than would be the case in most actual heating for hardening, in order to accentuate the tendencies of the gas toward carburization and decarburization. In instances where no staining occurred, the gain or loss of weight of the specimens is wholly due to carburization or decarburization. When staining did occur, in an active gas, the oxide coating was ordinarily of negligible weight compared to the weight change from gain or loss of C. Hence, weight changes may be taken as the measure of carburization or decarburization.

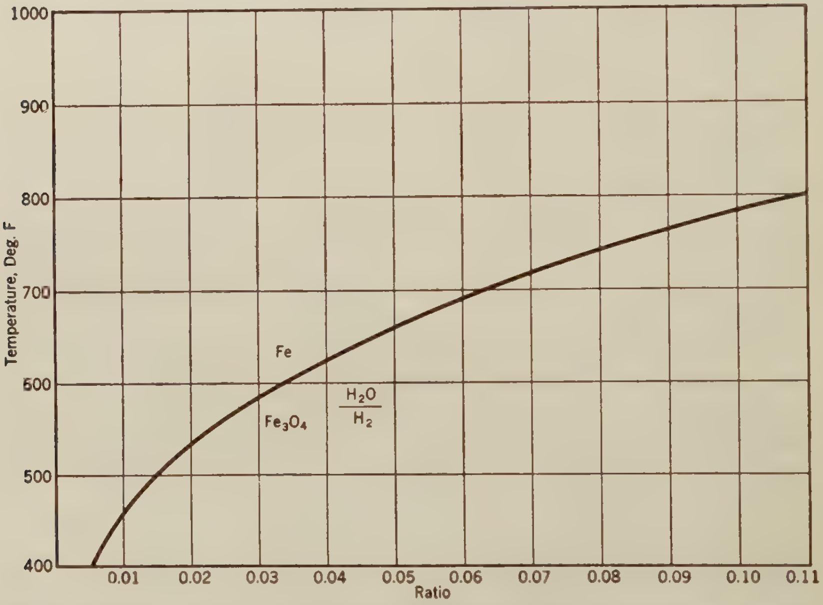


FIG. 12.—Ratios of $H_2O : H_2$ lying below the curve will scale or stain steel at the temperatures shown. Ratios above the curve will not stain, though decarburization will occur. (After Schenck.[23])

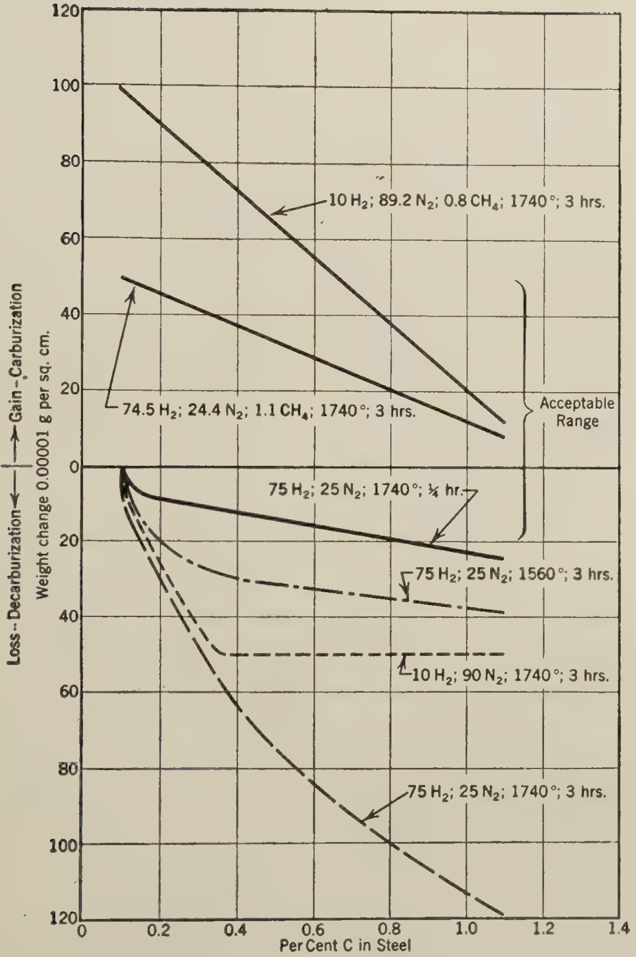


FIG. 13.—Effect of dry hydrogen, in varying dilution, at various temperatures and for various times. The curve for $10H_2$, 90% N_2 at 1740°, 1 hr., is not plotted, since it falls practically upon the zero line. The $75H_2$, 25% N_2 gas was cracked anhydrous ammonia. This shows that dry H_2 is decarburizing, but that decarburization is less as the time, the temperature, the percentage of H_2, or the C content of the steel is decreased. The samples were not stained. The upper curves show that the decarburizing tendency can be counterbalanced by addition of CH_4. By using some definite, still smaller amount of CH_4 the gas could be made neutral to steel of any C content. (Slowter and Gonser.[25])

Assumption as to Acceptable Range. By microscopic examination it was found that it required a weight increase of 0.00010 gram/cm.^{2} to produce detectable carburization or an equal loss for detectable decarburization, the weight changes being more sensitive than the microscope. It is calculated that a weight change of ± 0.00020 gram/cm.^{2} in the 3-hr. test would correspond to an entirely negligible action in the shorter time of ordinary heating for hardening. In a full spheroidizing anneal the time might be much longer but the temperature would be much lower, so that one would expect that an atmosphere that produced only ± 0.00020 change in the test would probably serve for spheroidizing. However, if the data are to be applied to a much longer exposure at the test temperature, as might occur in the case of heavy die blocks, one might consider that a change of ± 0.00005 to 0.00010 would indicate detectable carburization or decarburization. The curves here shown are average ones, to indicate trends, without plotting individual points. It is seldom necessary to quench steels of less than 0.20% C. Since a small amount of carburization can ordinarily be borne, a gas that shows less than +50 or −20 units weight change on steels of 0.20% C or over might ordinarily be considered satisfactory for commercial hardening of such steels as give results in this test within that range.

Effects of Time and Temperature. The effects of changes in testing time, temperature, etc., are brought out in Fig. 13, which shows that H_2 has a decarburizing action, decreased as the H_2 is diluted with N_2, as the temperature is lowered, as the time is decreased, and as the C content of the steel is decreased. The decarburization can be offset and the gas made carburizing by very small additions of CH_4. Figure 14, from Koebel,[26] shows the effects of these variables (other than dilution) with CO free from H_2O or CO_2 and brings out the fact that CO itself is a relatively inert gas notwithstanding that it is usually considered to be a carburizing gas. Figure 15 shows that the already small effect of the 34 CO : 66 N_2 gas is decreased as the dilution by N_2 is increased. The 5 CO : 95 N_2 curve lies practically on the zero axis. Such a gas, were it commercially available, would be suitable for steels of any C content, subject to the limitation that steels high in Cr will stain in CO.

Effect of Hydrogen Additions to Carbon Monoxide. If commercial heating for hardening were carried out in H_2 : CH_4 : N_2 or in CO : N_2 without any contamination by other gases, the problem would be simple. The commercial sources of CO usually involve contamination, so that purification may be required. In this connection the effect of mixtures of CO and H_2 may first be considered. Figure 16 shows that

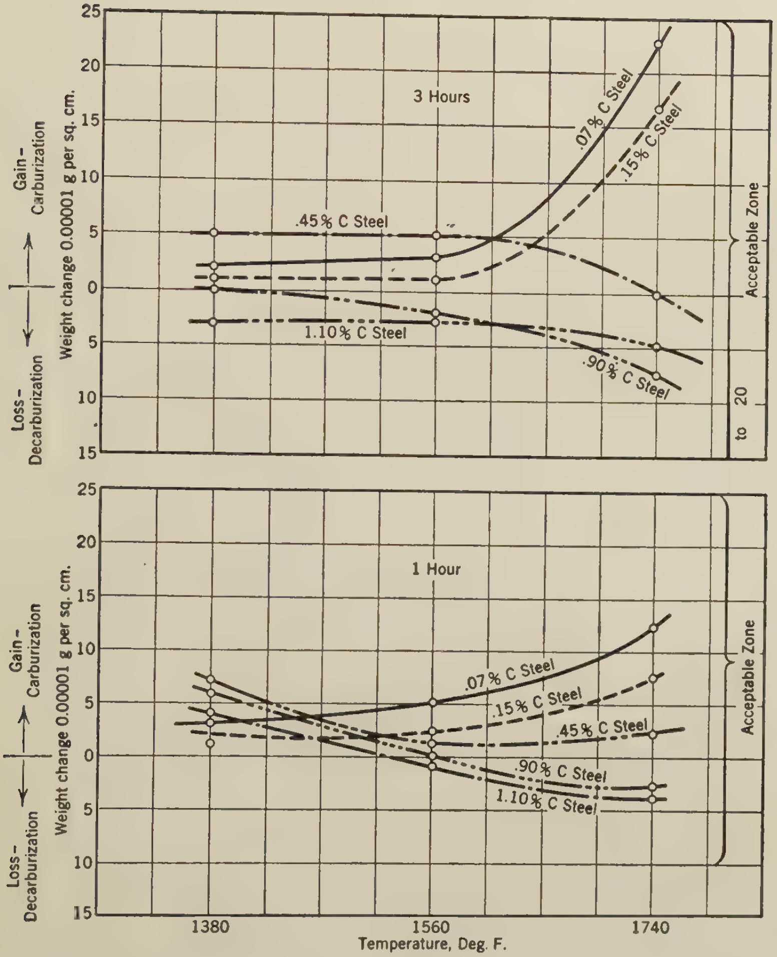


FIG. 14.—Effect of dry 34CO, 66% N_2 gas on steels of various C contents for 3 hr. and 1 hr. In the absence of CO_2 or H_2O the carburizing and decarburizing effects of CO are very small. The samples were not stained. In the lower graph the "acceptable range" has been halved because of the shorter time. (Koebel.[26])

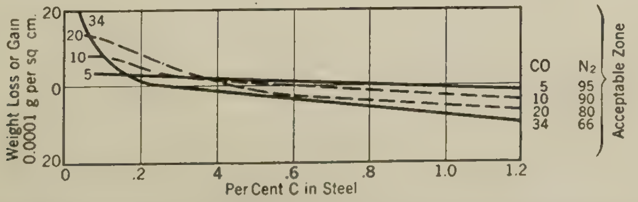


FIG. 15.—Action of dry CO in various CO : N_2 ratios. Steels exposed to the gases 3 hr. at 1740°. Samples not stained. (Slowter and Gonser.[25])

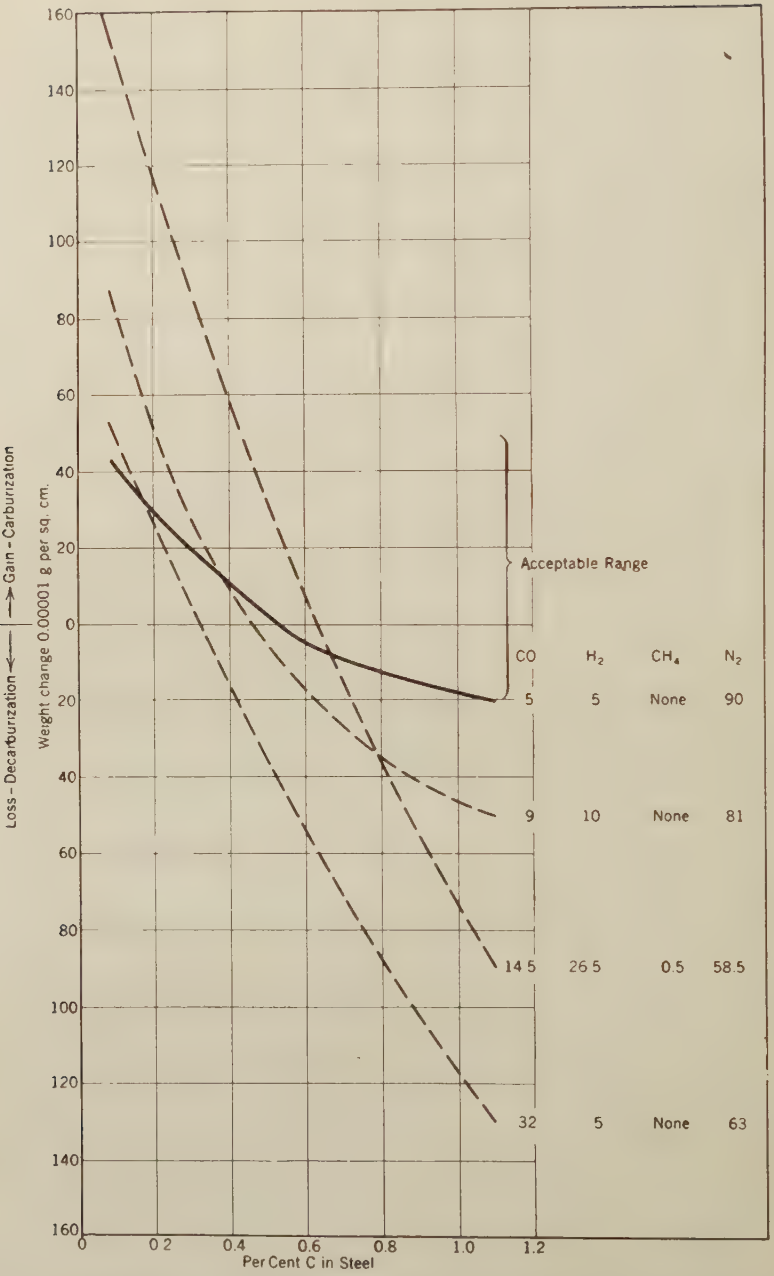
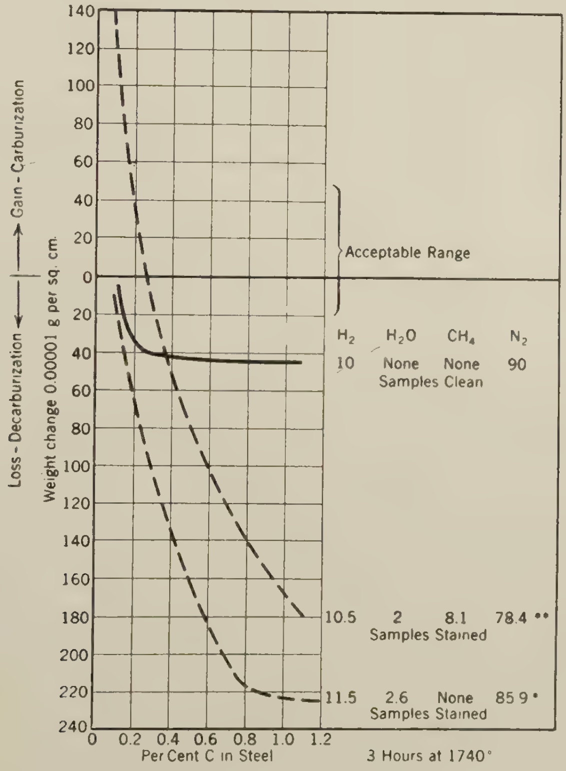


FIG. 16.—Effect of dry, CO_2-free mixtures of CO, H_2, and N_2 for 3 hr. at 1740°. The mixture of CO and H_2 is more strongly carburizing than CO alone, and more strongly decarburizing than H_2 alone. (The decarburizing effect can be counteracted by proper addition of CH_4.) Samples not stained. (Slowter and Gonser.[25])

the mixture is a more active carburizer to low-C steels than CO alone and a more active decarburizer to high-C steels than H_2 alone. However, suitable additions of CH_4 would serve to make the mixture carburizing, or neutral, as desired, to steel or any one given C content, as will be shown later.

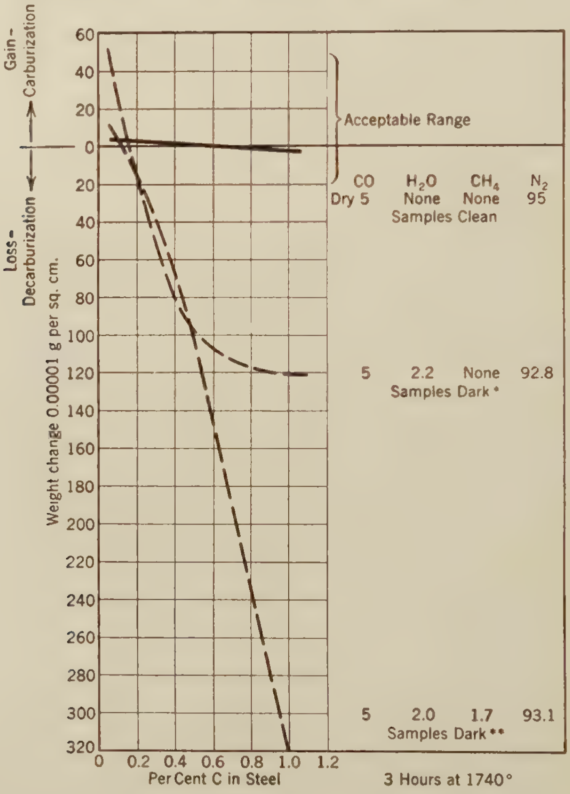
Effect of Water Vapor with Hydrogen. A usual contaminant of a commercial gas is water vapor. The effect of moisture in H_2 is brought out in Fig. 17. The presence, with 11.5% H_2, of 2.6% H_2O, an amount that would be held in the gas at 72°, vastly increased the decarburization. The H_2O reacted with the C of the steel, producing CO + H_2. The steel became stained on cooling, though, as Fig. 11 indicates, the gas was not oxidizing at the test temperature. When a large amount of CH_4 was added to H_2 carrying H_2O, in an attempt to oppose the



	H_2	H_2O	CH_4	CO	N_2
* Exit gas	12.9	1.3	—	0.8	85.0
** Exit gas	17.5	not det.	5.5	2.2	bal.

FIG. 17.—Effect of dry H_2, of H_2 with H_2O, and of the addition of a large amount of CH_4 to $H_2 + H_2O$. (Slowter and Gonser.[25])

decarburizing action of the wet H_2, the gas became carburizing to the 0.05% C steel; but its carburization went only to 0.25% C since the gas is neutral to steel of that composition. The H_2O and CH_4, as well as the C of the steel, reacted, as is shown by the composition of the exit gas. These samples were stained, also. Sykes[19] has studied this

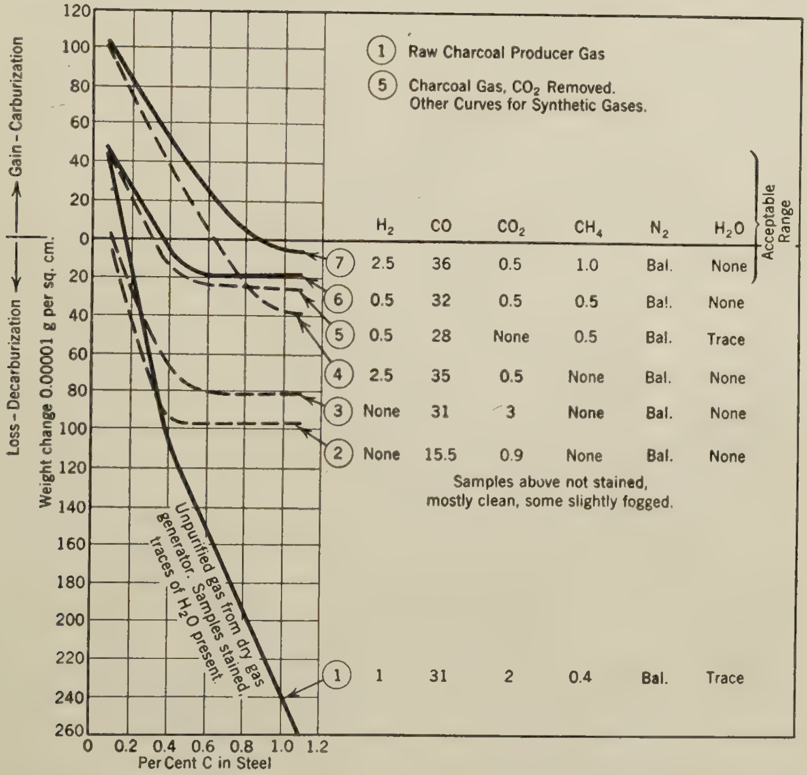


	H_2	H_2O	CH_4	CO	CO_2	N_2
* Exit gas	1.2	not det.	—	5.5	0.1	bal.
** Exit gas	1.8	0.4	0.9	5.6	0.3	bal.

FIG. 18.—Effect of H_2O in 5% CO, and of addition of CH_4 to wet CO. (Slowter and Gonser.[18])

problem. He found that at 1650° a gas of 99.5 H_2, 0.5 CH_4 was neutral to 0.18% C steel, but when 0.6% H_2O was present the CH_4 had to be raised to 2.5% to be neutral to the 0.18% C steel. The balancing of wet H_2 with CH_4 to prevent decarburization of medium- and high-C steel will be a difficult, if not an impossible, task. At temperatures necessary for hardening, the O in H_2O is so reactive that H_2 must be very thoroughly dried in order that the tendency toward decarburization may be kept in check, and to make the balancing by CH_4 effective. Marshall[27] considered that CH_4 was inactive, i.e., a mere diluent, in his work on bright annealing at 1200°, but CH_4 is not inert at hardening temperatures. Austin[47] has also studied the relations of H_2 and H_2O.

Effect of Water Vapor with Carbon Monoxide. The effect of water vapor with CO is shown, for 5% CO, in Fig. 18. Bad decarburization appeared and, as the exit gas showed, the gas was reactive. On adding CH_4 to the wet CO, the decarburization of the high-C steels was increased instead of decreased. The H_2O reacted with the CH_4 added as well as with the C of the steels.



	H_2	CO	CO_2	CH_4	N_2	H_2O
⑦	2.5	36	0.5	1.0	Bal.	None
⑥	0.5	32	0.5	0.5	Bal.	None
⑤	0.5	28	None	0.5	Bal.	Trace
④	2.5	35	0.5	None	Bal.	None
③	None	31	3	None	Bal.	None
②	None	15.5	0.9	None	Bal.	None
①	1	31	2	0.4	Bal.	Trace

Samples above not stained, mostly clean, some slightly fogged.

Unpurified gas from dry gas generator. Samples stained, traces of H_2O present.

FIG. 19.—Charcoal gas. Effect of varying CO_2, H_2, and CH_4 in dry CO : N_2 gas. Samples heated 3 hr. at 1740°. (Slowter and Gonser.[25, 35])

Whereas a gas higher in CO might not be quite so sensitive to H_2O, it is plain that, to avoid severe decarburization of medium- and high-C steels in heating for quenching, moisture would have to be kept out of the CO: N_2 gases in general.

Effect of Carbon Dioxide. The commercial gases containing CO almost invariably contain considerable quantities of CO_2 as well. The effect of CO : CO_2 mixtures has been shown in Fig. 10, from which the tolerance of CO for CO_2 may be seen. If we remember that it is the ratio, rather than the absolute percentage, that counts, since, as Fig. 19 shows, a gas of some 30–35% CO must not only be dried but must also have its CO_2 removed down to about 0.5%, it appears that one of around 10% CO would need to be freed from CO_2 down to around 0.2% in order to protect medium- and high-C steels from decarburization in heating for hardening.

Commercial Sources of Protective Gas. So far, the gases have been discussed without much reference to their commercial sources. Dry 75 H_2 : 25 N_2 can be obtained by cracking anhydrous NH_3 (Fig. 20). By adding CH_4 (as natural gas) or using some other gaseous hydrocarbon, or by bubbling the cracked NH_3, or a part of it, through a suitable liquid hydrocarbon, a suitable protective gas may be obtained. The cost may be reduced by burning the 75 H_2 25 N_2 gas with air, a deficiency of which adds more N_2, and drying the resulting H_2 : N_2 gas (which may, for example, be 10 H_2 : 90 N_2) to free it from the H_2O formed by combustion of the remainder of the H_2. A 5 to 10 H_2 : 95 to 90 N_2 gas has been made commercially by another process described by Kinzel.[28] In this, a fuel gas is partly burnt with air, treated with steam over a catalyst to burn the C of the fuel to CO_2 instead of CO,

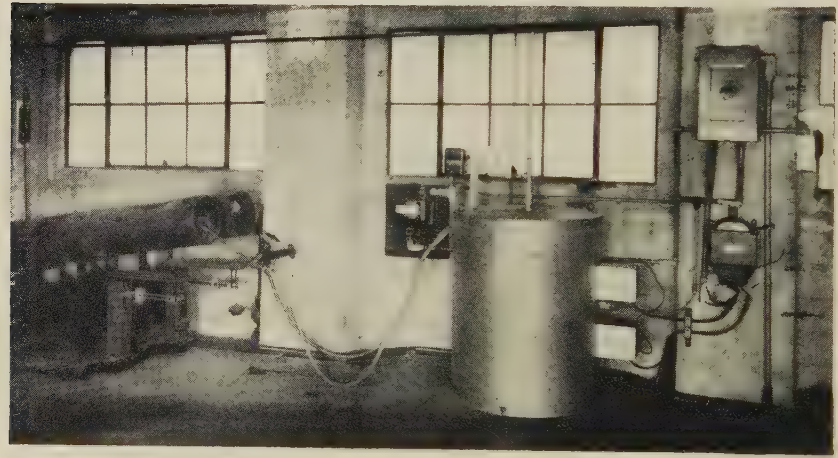


FIG. 20.—Equipment for cracking ammonia. Anhydrous ammonia from tanks is led to an electrically heated dissociator in which it is cracked. The apparatus is very simple. (Courtesy General Electric Co.)

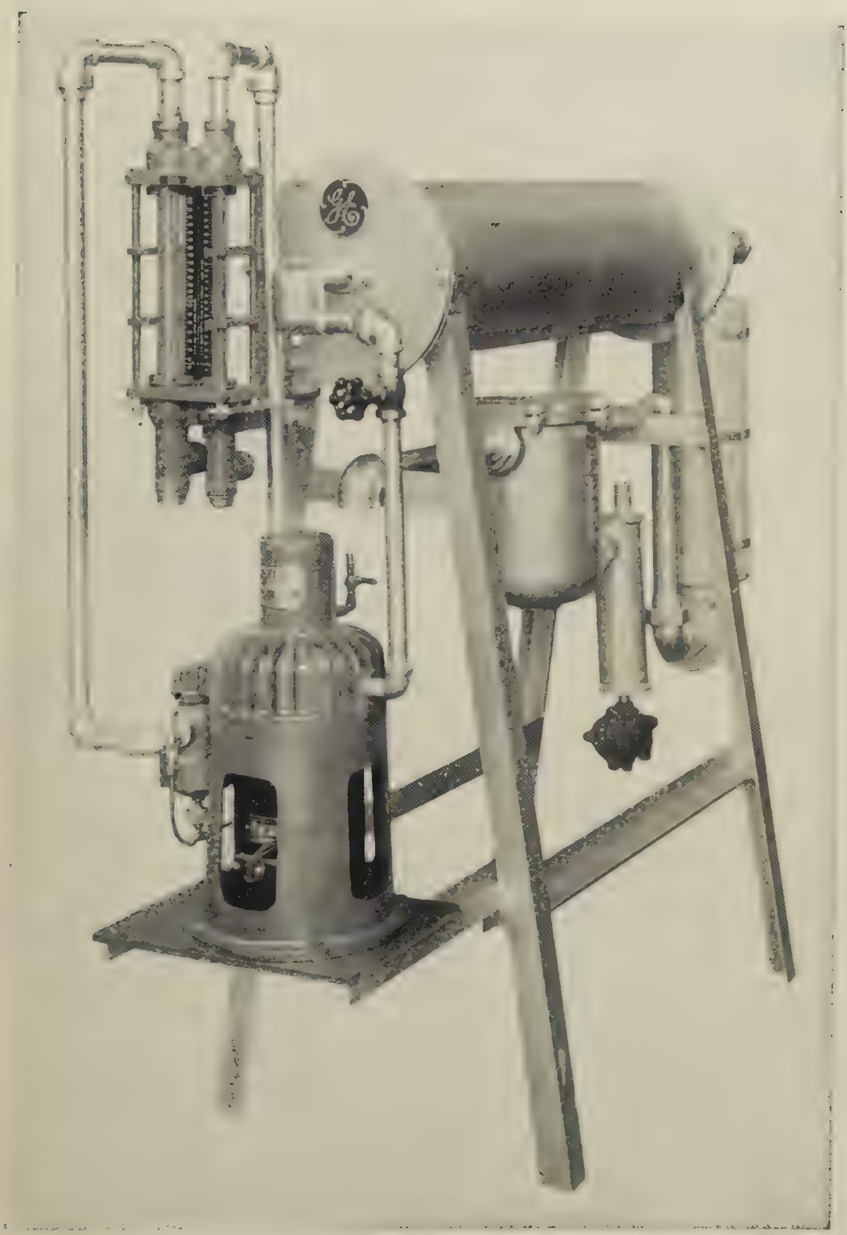


FIG. 21.—Small-scale generator equipment in which definite ratios of gas and air are mixed and burned at regulated temperature. The resultant gas is cooled to condense part of the moisture. (Courtesy General Electric Co.)

the CO_2 then chemically removed and the gas dried. A small amount of CH_4 is probably present in such gas. The 10 H_2 : 90 N_2 : traces of CH_4 mixture, has been used in annealing high-C steel wire in some very large installations. Straight cracked NH_3 costs about $3.50 per M for the resultant 75 H_2 : 25 N_2 gas. One of 10 H_2 : 90 N, produced from NH_3, costs upward of $2.00 per M. On a very large scale it has been calculated that the 5 to 10 H_2 : 95 to 90 N_2 gas can be produced by the method of Kinzel for $0.20 per M, but this would hold only for very large installations because of the high first cost of the equipment.

A gas, which, if it could be produced of theoretical purity, would consist of 34 CO: 66 N_2, can be made from charcoal, or other pure C and air, in a "dry gas producer" (charcoal gas producer) at perhaps $0.50 per M. Charcoal gas is in commercial use. Robiette [29] discusses its recent use in England; it was used long ago in Sweden [30]; Cone,[31] Holler,[32] and Gould and Toole [33] record its use for other purposes; and Dow [34] mentions present use in this country for bright hardening and, with the addition of 1 to 2% propane, for producing a very thin case on Buick transmission gears while heating for hardening.

Gas from Dry Gas Producers: "Charcoal Gas." If thoroughly dry air and charcoal are used, if the temperature is held above 1800° and if time is allowed for attaining equilibrium, the gas should contain about 33.5% CO, 0.5% CO_2, remainder N_2. Under usual operating conditions it will contain 25 to 30% CO, 1 to 3% CO_2, 1 to 3% H_2, and a little CH_4. By scrubbing out CO_2 with caustic soda (NaOH) or one of the ethanolamines (which are often preferred because they can be regenerated, i.e., the CO_2 they have absorbed can be driven off by heating), the CO_2 can be removed. It is difficult to remove the last traces, and in the purification of any gas mixture from CO_2 it is very likely that in actual operation there will be periods when 0.25 to perhaps 0.5% CO_2 will be left in the gas. As Fig. 10 shows, a CO : CO_2 ratio of 40 : 1 will not decarburize steel of any C content, so, with 0.5% CO_2 and 20% or more CO, the gas should serve for prevention of decarburization of high-C steels. In the bright normalizing of *low*-C steels a 10 : 1 ratio would suffice, i.e., a 30% CO gas containing 3% CO_2 should be neutral. Hence un-scrubbed charcoal gas might serve for normalizing of low-C steel. (See Fig. 19). At bright annealing temperatures still higher CO_2 could be tolerated.

From the point of view of decarburization, the presence of some H_2 must be taken into account. As was shown in the earlier discussion of hydrogen (Fig. 13), even dry H_2 has some decarburizing tendency. In the absence of CO_2, the CO : N_2 mixture of a completely scrubbed and dry charcoal gas will be the more carburizing to low C and the more decarburizing to high C, the higher the H_2. In the presence of CO_2, even though the CO : CO_2 ratio by itself might indicate a neutral gas, addition of H_2 or CH_4 allows the water-gas reaction:

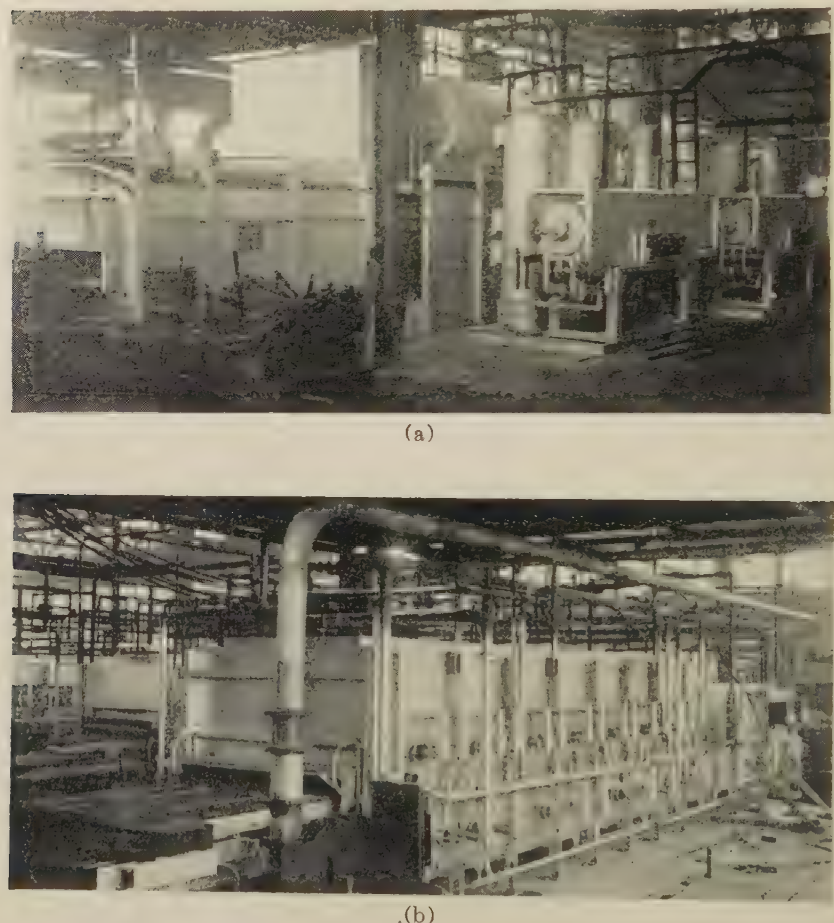
$$H_2 + CO_2 \rightleftarrows CO + H_2O$$
$$CH_4 + 3CO_2 \rightleftarrows 4CO + 2H_2O$$

to take place, with the production of moisture. The gas system then becomes more complex, the bad effect of H_2O appears, and the attempt to force the gas over to the carburizing side by addition of hydrocarbon is opposed and tends to be defeated by the decarburizing action of the moisture.

Slowter and Gonser[35] studied complex gases of this type. Figure 19, curve 1, shows the results on a gas direct from a dry gas producer, not purified. The entrance gas was dry, since H_2O is decomposed to form H_2 + CO with the C in the dry gas producer, but H_2O is later formed as the CO_2 reacts. When CO_2 is removed, curve 5, the tendency toward decarburization is vastly decreased. With 2.5 H_2, 35 CO, 0.5 CO_2, curve 4, the gas is more carburizing to low-C and more decarburizing to high-C steels than without the H_2 (compare curve 6); but, by adding a small amount of CH_4, curve 7, such gas can be made practically neutral to steel of 0.60 — 1.10% C, and mildly carburizing to those of higher C. The gas was slightly reactive owing to the CO_2, but the steels were only fogged rather than stained. Consideration of the cost of charcoal and the operating cost of a dry gas producer indicates that the gas should cost something of the order of \$0.50 per M. No actual cost figures are available.

"Generator Gas." A still cheaper gas can be made by incomplete combustion of a fuel gas with air. With natural gas and a 1 : 6 gas : air ratio the resultant gas is figured at \$0.10 to \$0.20 per M. Such a gas, with more or less complete drying, has been widely used for low-temperature bright annealing. As Thum[24] points out and Fig. 10 shows, almost any fuel gas, burnt so as to be surely free from O_2 and containing not more than half as much CO_2 as CO, if properly dried, is suitable for bright annealing of low-C steel. The reactivity of H_2 or CH_4 with CO_2 is so low at 1200–1250° that harmful amounts of H_2O are not formed in the gas. The incoming gas, however, should be thoroughly freed from H_2O.

The gas is usually generated in special equipment, such as is shown in Figs. 7, 21 and 22, in which a higher gas : air ratio is used than would be used in burning the gas for heating. However, flue gas may be used if it is of suitable character. A few gas-fired furnaces have been built in which a part of the flue gas is diverted, passed through a drying system, and recirculated to the furnace chamber, instead of preparing the atmosphere in a separate generator. These have been used chiefly for nonferrous annealing and for clean annealing of low-C steels.


(a)

(b)

Fig. 22.—Large-scale equipment for regulated combustion of fuel gas and air is shown at the right in (a). The resulting gas is cooled to separate most of the moisture and led into the furnace. In (b) the other end of the furnace is shown. The furnace is at the Detroit Steel Products Co. (Courtesy Surface Combustion Co.)

Dow [34] cites a large continuous gas carburizing furnace in which part of the flue gas is recirculated to the charge, after enrichment with natural gas. Obviously, great care must be taken to avoid an excess of air and consequent presence of free O_2, and the fuel : air ratio must be such that the resultant CO : CO_2 ratio will not cause staining. Most furnace makers prefer to prepare the controlled atmosphere in a separate small unit in which the fuel : air ratio is higher than in the burners for heating.

It is recognized, and has been brought out by Heyn,[36] that unpurified generator gas, even though dried, is not suitable for prevention of decarburization in medium- and high-C steels.

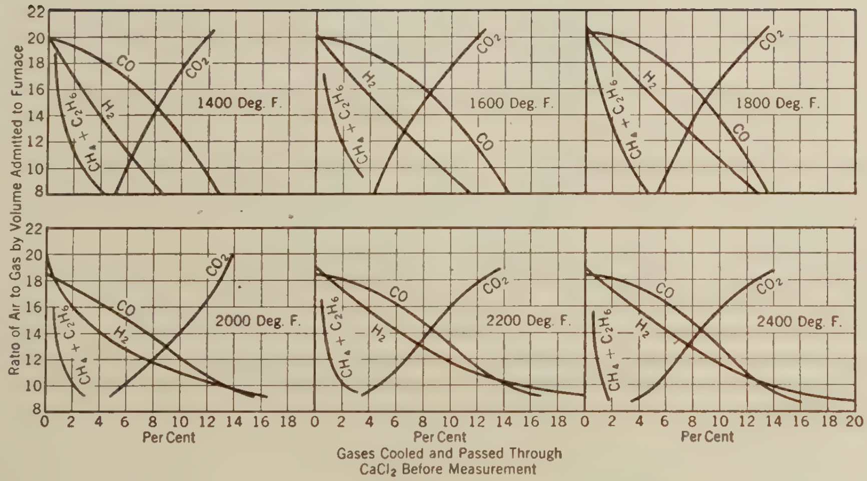
Composition of Generator Gas. The exact composition of the gas will depend on the fuel gas used, the gas : air ratio, and the temperature at which the gas is burnt. Figure 23 shows some of the variations possible. Marshall [27] and Krogh [37] have given similar curves for city gas. The gas : air ratio must be kept high enough so there will be no danger of getting any free O_2. The ratio is usually adjusted to give a gas of around 5–7 CO_2, 9–10 CO, 9–10 H_2, 0.5–2% CH_4, remainder N_2. Compare Owen [38] and Thum.[24] This may be considered a typical generator gas.

The H_2O content depends on the cooling or drying of the gas. The raw gas, hot, and given no opportunity to deposit its water vapor, will contain 15 to 20% water vapor. If cooled to 46°, as by a cold water spray, 1% H_2O will be retained (see Fig. 8).

Effect of Raw Gas. Some of the early "controlled atmosphere" furnaces led the fuel gas and air directly into the furnace chamber without any opportunity for separation of H_2O. The effect of this is shown in the curve (1) of Fig. 24, with extremely bad decarburization. Even at the lower temperature of this test, both decarburization and scaling occurred. In another test, not plotted, at a still lower temperature, 1560°, and for only 15 min., all the steels were scaled and the 0.45% and higher C steels were appreciably decarburized.

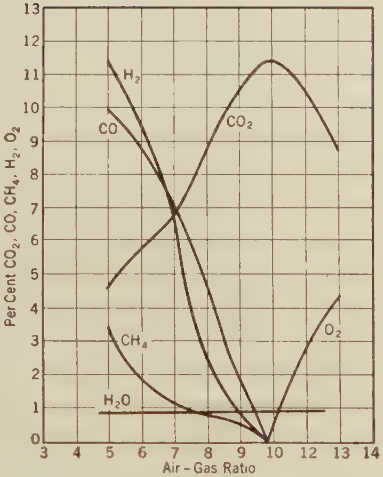
Effect of Increased Methane. Altering the gas : air ratio to leave very high CH_4, as in curve (4) of Fig. 24, was unavailing. Merely drying the entrance gas, as in curve (2) was also unavailing. Operating at 1650° or 1740° with an entrance gas of 5.2 CO_2, 9.2 CO, 9.3 H_2, 0.9% CH_4, no H_2O, the exit gas contained about 2.9 CO_2, 12.5 CO, 8.6 H_2, 0.85% CH_4, H_2O not determined. The H_2 and CH_4 reacted with CO_2, more CO was formed, i.e., the gas was active, and H_2O was being formed by the reaction $CO_2 + H_2 \rightleftarrows H_2O + CO$. This provided enough H_2O to make the gas very strongly decarburizing.

When the CO_2 was removed, as in the upper curves, this reaction did not occur. The gas became more carburizing to low-C steels as the CO_2 was progressively removed. When the CO_2 was brought low enough and the CH_4 made high enough, decarburization was prevented on all the steels.



(a)

FIG. 23. (a).—Change in composition of products of combustion of propane with varying air ratios and temperatures of combustion. (Krogh.[51])



(b)

FIG. 23 (b).—Effect of variation in air ratio on the products of combustion of natural gas, after cooling to 40°. (Heyn.[36])

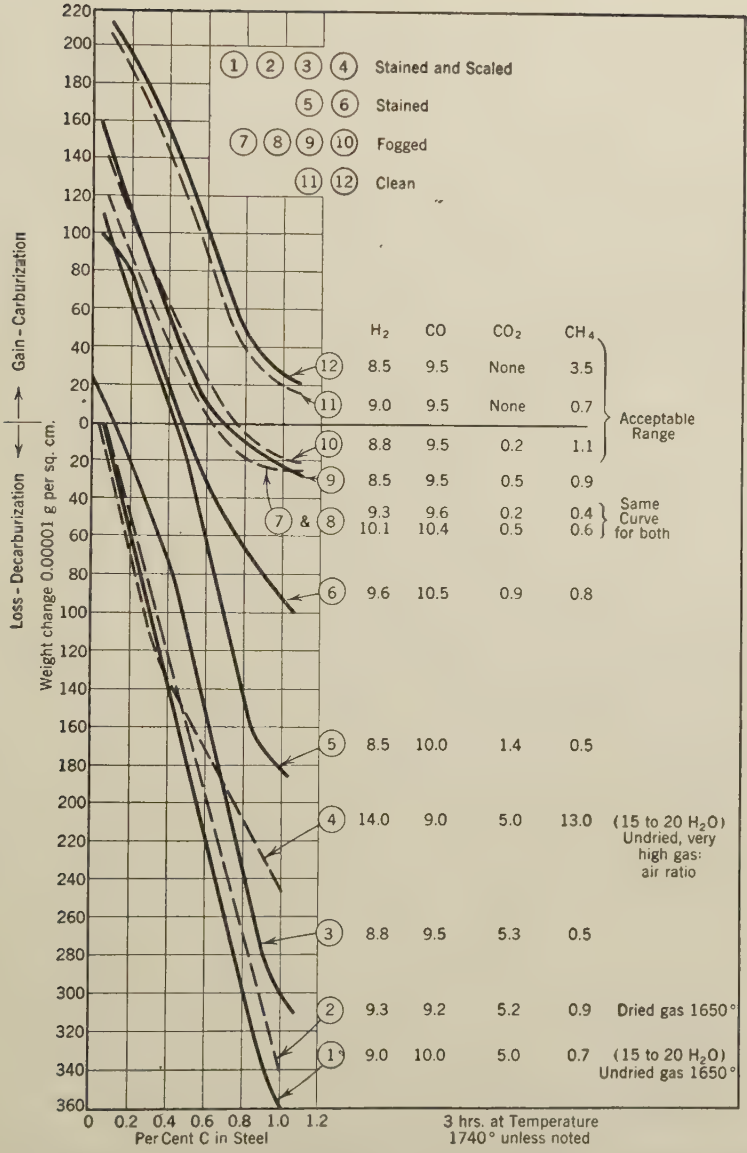


FIG. 24.—Effect of drying, of purification from CO_2, and of addition of CH_4, on the behavior of "generator gas." See text for composition of exit gases. (Slowter and Gonser.[18, 25])

Necessity for Purification. Thus it becomes evident that, if such a gas is to be used to avoid decarburization of high-C steels, the raw generator gas must be thoroughly dried, the bulk of the CO_2 removed, and a suitable amount of CH_4 introduced either by adjusting the gas : air ratio, so that the necessary amount of CH_4 is formed, or by adding enough to bring it up to the required amount.

As was mentioned in connection with charcoal gas, in the commercial purification of a gas from CO_2, it is difficult to obtain complete removal of the CO_2. Caustic towers, or some amine, like ethanolamine, will be used. With the latter, after the solution has absorbed so much CO_2 as to be inefficient in further absorption, the solution is regenerated by boiling. In practice the gas will not be switched from the spent absorber to the fresh one till the last possible minute, whether NaOH or an amine is used, and the removal of CO_2 will, at times, be somewhat incomplete.* The tolerance of the H_2 : CO mixture for CO_2 and the necessary amount of CH_4 must therefore be considered.

Gonser and Slowter [25] tried generator gas, undried, dried, and dried with CO_2 completely removed, on a series of C and alloy steels, using three temperatures instead of one, so as to secure data in the neighborhood of the temperature likely to be used in heating the different steels for hardening. Some of their data are given in the table below.

The generator gas contained 9–10 CO, 9–10 H_2, 0.5–1 CH_4, 5% CO_2, and, if used raw without drying, would also carry more than 15% moisture. The losses or gains in weight tabulated (see page 329) are given in the usual unit of 0.00001 gram/cm.^{2}

The decrease in the decarburizing tendency with drying and with removal of CO_2 is evident. Steel AA could be bright normalized in the dry, CO_2-free gas without staining, but not in the presence of the moisture or CO_2. A transformer steel of 0.03% C, 4% Si was not carburized at 1650° for 3 hr. in any of these atmospheres, but was stained unless the CO_2 was removed. In general the dry, CO_2-free gas is carburizing to everything but the very high C steel. The addition of a little more CH_4 would make it carburizing to that steel at the temperature used. At 1740° the dry CO_2-free gas was faintly carburizing to the high-C steels, steel D gained 16 units in 3 hr. and steel E, 10 units. Raising the CH_4 to 3.5% increased the carburization on these steels, at 1740° for 3 hr., to 38 and 24 units respectively.

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* A continuous system in which the amine is circulated, the CO_2 being boiled off in one part of the system, the purified amine then cooled and recirculated to the absorbent chamber, ensures constant activity of the absorbent. Such equipment is coming into use. More absorbent is required in such a system, but the gas is more consistently purified.

Steel		Undried Generator Gas	Dried Gas	Dried and CO_2 Removed	Temperature, Deg.
AA	0.07 C	+ 17 *	+ 1 *	+ 76 †	1740
G	0.19 C	− 43	− 24	+146	1650
H	0.44 C	− 91	− 4	+ 85	1470
D	0.87 C	−147	− 27	+ 94	1470
E	1.02 C	−389	−366	− 76	1650
S	C Si Mn 0.50 2.00 0.75	−104	− 99	+ 23	1650
O	C Cr V 0.45 0.95 0.18	−134	−105	+ 56	1650
J	C Cr Al Mo 0.23 1.58 1.24 0.20	+ 48 *	+ 37 *	+ 36‡	1740

* Stained. † Bright. ‡ Not stained, but showed "chromium fog."

It is of some interest to note that Si-Mn spring steel, which is very prone to decarburization when heated in air or in flue gas with some free O_2, and is appreciably decarburized in the raw or the dried generator gas, is not decarburized in the dry, CO_2-free gas. Also that the Al-Cr-Mo nitriding steel, which, as is mentioned in Chap. 13, must be free from decarburization in order to be satisfactorily nitrided, is mildly carburized at 1740° by the gas in any of the three conditions, as might be expected from its low C content. These examples indicate that, when using protective gases of this type, the degree of purification must be fitted to the particular steel and temperature.

Effect of Carbon Content of Steel. In this fitting, the C content of the steel is of supreme importance. In the diagrams from the work of Gonser and Slowter, a type of gas that is universally applicable to a wide variety of steels gives a curve that is quite flat and hugs the zero weight change line, while a type that must be adjusted to the particular C content of the steel crosses the zero line at a steep angle. The 5 to 35% CO mixtures with N_2, Fig. 15, are of the first, or fool proof type, the generator gas mixtures, Fig. 23, are of the second type. The charcoal gas, Fig. 19, properly purified, is more nearly fool proof than the generator gas.

Were gases of the first type more readily available, they would be much more convenient to use. If a gas is used whose degree of purity and whose necessity for balancing with CH_4 or other hydrocarbon are crucial for a given C content of the steel, commercial variations of C may affect the ersults. Moreover, as Gonser and Slowter found for two plain C steels of 0.40 and 0.41% C, steels that are closely similar in chemical composition may vary appreciably in ease of carburization or decarburization. Hence the charts shown herein cannot be taken as an exact guide in the case of a sensitive gas whose curve crosses the zero line at a large angle.

Active Gases Seek Equilibrium. Reverting to the necessity for knowledge of the degree of purification of generator gas from CO_2, it was shown above that the charcoal dry producer gas with 20 to 30% CO required the elimination of CO_2 down to about 0.5%. Since, in general, it is the ratio of a gas pair that counts instead of the actual percentage, it is reasonable to expect that in generator gas, with around 10% CO, the CO_2 elimination, for use in hardening without decarburization, should be carried further, presumably to somewhere around 0.25%. This may be brought out in another way. If the gas is neutral to the steel and if its own components do not react together at the operating temperature, the outlet gas will have the same composition as the inlet gas. Gonser and Slowter found, in tests at 1740°, passing the gas over steels ranging from 0.05 to 1.05% C, that the inlet and outlet gases from generator gas dried, and purified from CO_2 to various degrees, gave results like the following.

	CO_2	CO	H_2	CH_4
In	1.4	10.0	8.5	0.5
Out	0.3	11.0	9.1	0.5
In	1.0	10.0	9.5	0.9
Out	0.3	11.0	10.0	0.9
In	1.0	10.0	8.3	0.5
Out	0.2	10.5	9.7	0.5
In	0.9	10.5	9.6	0.8
Out	0.4	10.8	10.1	0.6
In	0.5	10.4	10.1	0.6
Out	0.2	10.8	10.5	0.7
In	0.5	9.5	8.5	0.9
Out	0.2	10.1	9.2	0.9
In	0.2	9.6	9.3	0.4
Out	0.2	9.6	9.3	0.4
In	0.2	9.5	8.8	1.1
Out	0.15	9.6	9.5	0.9
In	None	9.5	9.0	0.7
Out	None	9.5	9.0	0.7

It will be noted that with inlet gas ranging from about 1.5 to 0.25% CO_2 the outlet gas runs about 0.25% CO_2. This is strong evidence that the gas is reactive and seeks to attain an equilibrium. Inlet gas of the equilibrium composition should be more nearly inert to the steel than a gas that is reactive, and it ought to be easier to correct such a gas by regulated CH_4 additions. Leading a prepared gas with active constituents over steel chips of the composition of the work to be treated, and at its temperature, in a separate reaction chamber before it meets the work, or even through a hot steel tube inserted in the furnace, gives the gas some opportunity to reach equilibrium. A scheme of this sort has been mentioned by Lewis.[39]

Purification and Adjustment of Generator Gas for Hardening. The effect of removing CO_2 is shown in Fig. 24. If the assumption that +50 to −20 units is the acceptable range be correct, it appears that a degree of purification from CO_2 that would represent very good commercial practice would be required to make dry generator gas sufficiently neutral for use of very high-C steel.

It will be noted that in the operation of the generator at the given setting for the natural gas : air ratio used, the unburnt CH_4 varies from about 0.5 to 1% and the variation in CH_4 should not be neglected. To find whether a generator gas from which CO_2 had been incompletely removed could be compensated by adding CH_4, Gonser and Slowter [35] made the experiments shown in Fig. 20 which indicate that, with a gas of 10% CO, and 10% H_2, the CO_2 should be removed down to about 0.2% and the CH_4 held at about 1% in order to make the gas satisfactory for use with steels of more that 0.40% C. With CO_2 ranging from 0.5 to 2%, the CH_4 content is quite critical. Slight variations materially alter the behavior of the gas. Gonser and Slowter pointed out that with 10 CO, 10 H_2, 0.2–0.5% CO_2, the amount of CH_4 (in volume percentage) required to make the gas neutral to a steel is roughly the same as the percentage of C in the steel. With CO_2 all removed, the CH_4 content is not critical, since, as curves 11 and 12 in Fig. 24 show, 0.7% and 3.5% CH_4 give similar results. An excess would naturally be avoided to prevent deposition of soot.

Charcoal Gas Requires Less Complete Purification. From these data, compared with those of Fig. 19 for dry charcoal producer gas, one wonders whether the lower cost of the generator gas over the charcoal gas may not be illusory when high-C steels are being heated for hardening, provided practical charcoal producers are designed. Holler [32] shows that 6½ to 7 lb. charcoal will provide 1000 cu. ft. of gas. The charcoal gas, with its higher CO, can tolerate more CO_2, and hence does not demand so much care in the control of the process of scrubbing out CO_2. Improvement in methods of CO_2 removal would be welcome in respect to either charcoal gas or generator gas. It has been suggested that after the bulk of the CO_2 has been removed, say down to 0.5%

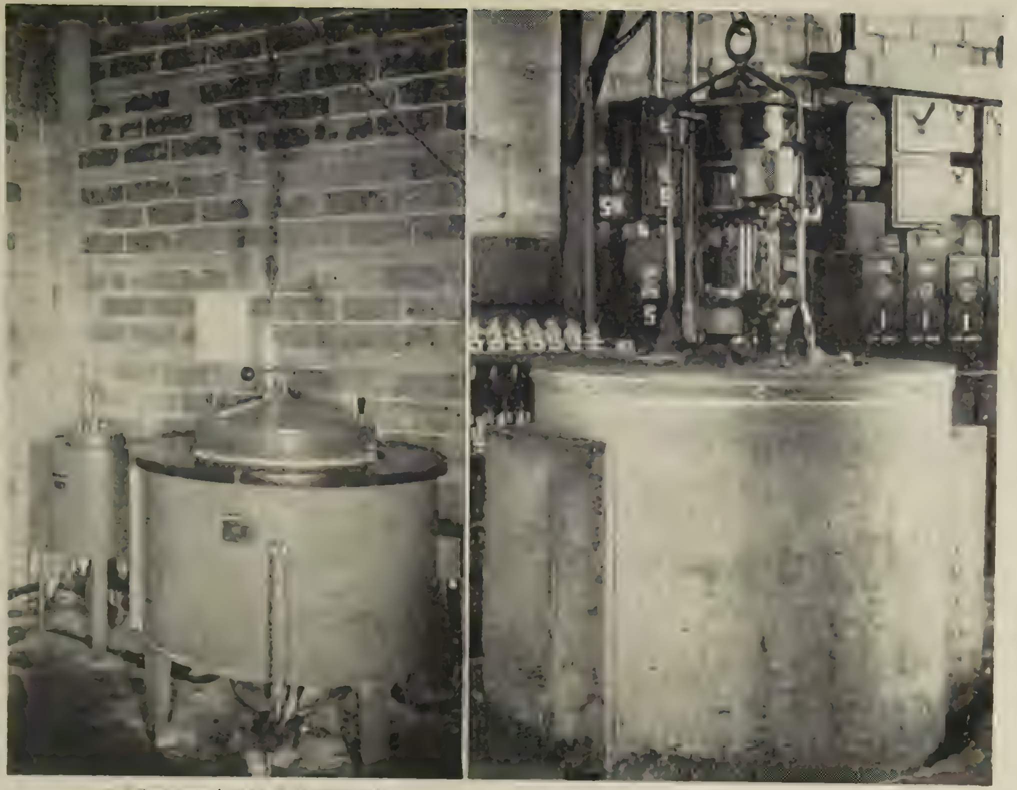
CO_2, mixing a little NH_3 gas with it and scrubbing with water, preferably compressing the gas at the same time or scrubbing with NaOH,[40] would eliminate all but traces of CO_2; but no record has been found of such practice applied to protective gases of the type under discussion.

Since decarburization is ordinarily more to be avoided on high-C steels than a slight increase in surface C, it would be reasonable to use the well-known gas-carburizing methods in heating for hardening, since the time at temperature is ordinarily relatively short.

Oil Vapor. Oil vapor heated in a separate unit so as to make it deposit its soot there and be cracked into H_2 and CH_4 before going to the furnace itself or, where some soot deposition on the work will do no harm, oil dropped directly into the furnace is often employed when a carburized surface is acceptable. The procedure is the same as in actual gas carburizing, but the shorter time and lower temperature for hardening give only a thin carburized layer. Obviously the other gas-carburizing atmospheres commercially available might be used, but they should be chosen to minimize the deposition of soot. (See Fig. 25).

Gas from Reaction of Air with Cast-Iron Borings. Steel may be packed in cast-iron borings and be heated for quenching without scaling or appreciable decarburization if conditions are favorable. The O_2 of the air is mostly used up in oxidizing the iron of the borings, the remainder reacts with the graphite and C in the borings, and any moisture in the air has its O combined either with Fe or C and its H freed. The gas reaching the work is then N_2 carrying a little CO and H_2. The effect of such a gas, prepared outside and led into the furnace chamber, is shown in Fig. 26. The gas from cast-iron borings behaved about like the $10\ H_2 : 90\ N_2$ gas of Fig. 17. While the effect is slightly decarburizing, when the conditions are not too exacting, the simple expedient of packing work in cast-iron borings should be effective.

Gas from Charcoal Packing or Carbon Blocks. Since pure charcoal, coke, or graphite, without energizer, has only a mildly carburizing action on high–C steels, as is shown by the long time necessary to bring the surface C content up above eutectoid composition on steels so carburized, packing in some form of non-activated C or in sand mixed with C is often employed. Hankins and Becker[41] have discussed this in some detail in reference to spring steels. Grossmann[42] and others point out that during slow cooling of the charge in the packing, decarburization can occur. Curran and Williams[43] report that 1.15% C steel is strongly decarburized in 4 hr. at 1425° and slightly decarburized in 2 hr. at 1675° in non-activated charcoal, but that by adding Na_2CO_3 or $BaCO_3$ the decarburization can be avoided. They suggest adjusting the amount of activating carbonate to make



(a) (Courtesy Leeds & Northrup)

(b) (Courtesy Hevi Duty Electric Co.)

In (a), oil from the container upper left, drips into the cracking unit, lower left. The resulting gas is led to the furnace.

In (b), oil drips onto the blades of the fan in the top of the heating chamber. The oil cracks, and the resulting gas passes over the work.

Fig. 25.—Carburizing furnaces. They are also used for hardening. Oil is used to supply the carburizing gas.

the packing neutral to high-C steels. This is all in accord with Fig. 10.

Carbon blocks in proximity to the work in a muffle furnace are commercially used to produce some approach to a pure CO : N_2 atmosphere. The contamination with CO_2 will depend on the temperature and the rate of air flow. Electric furnaces with granular C or graphite resistors or heated by resistors composed of stacks of C plates, operating in air, also tend to produce atmospheres high in CO, and higher than when the C is heated only to the temperature of the work, since the resistors are much hotter and this raises the ratio of CO to CO_2.

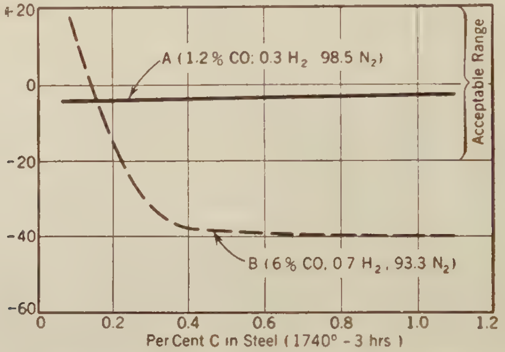


Fig. 26.—Effect of gases made to simulate packing in steel or cast-iron chips. A was made by passing air over hot steel borings plus a very little C. B was made by passing air over cast-iron borings and scrubbing out any CO_2. B was slightly more reactive than the 5CO, $5H_2$ synthetic gas of Fig. 16, perhaps because of S in the cast iron. Samples were bright in both A and B. (Slowter and Gonser.[18])

Effect on Heat-resistant Alloys. A factor in overall cost is the effect of the protective gas upon the Ni-Cr and similar alloys used for resistors, radiant tubes, muffles, or conveyor parts in continuous furnaces.[44] Some atmospheres at some temperatures are deleterious to these alloys. The effect is especially to be avoided in the case of resistors, since, if the gas attack is of a localized type, the resistors will burn out at such a spot. Sulphur is generally considered harmful from this point of view. It is usually held that strongly carburizing gases should be avoided for long life of such alloys, though alloy carburizing boxes are in general use. An oxidizing atmosphere in which a tightly adhering oxide is produced on these alloys would be most suitable from this one standpoint.

The problem is being actively studied but, because of the long life demanded in service, the evaluation by rapid tests of deterioration that is so slow in actual service is a difficult matter.

It would appear that a gas that was neutral or slightly carburizing to the steels being heat treated, and at the same time slightly oxidizing to the heat-resistant alloys, could be selected when further information becomes available.

Guarding against Impurities. Moisture. The drying of any of these gases deserves a work of separate comment.

The moisture content is seldom reported in percentage in discussions of protective gases, since the usual gas analyses do not include its determination. It is more commonly expressed in terms of the temperature to which the gas has been cooled, since there is a characteristic dew point for each temperature and the gas will deposit its contained moisture on cooling till that amount has been reached which corresponds to that dew point. Hotchkiss[45] tabulates dew points in terms of grains H_2O/cu.ft. and volume per cent. At 65° the gas holds about 2% H_2O, at 45° 1%, and at 0° only 0.15%. These data are given in Fig. 8. The determination of moisture content is conveniently done by observing the dew point of the gas. Apparatus for this is on the market.[46]

Thus, if the gas is cooled by a water spray, the exit gas contains moisture corresponding to the temperature of the water. Water may contain air in solution and the water spray may thus introduce traces of free O_2 which will stain the work. Passage of the incoming gas through hot iron or steel borings will take out the free O_2. Cooling coils are often operated at 35°, so that the condensed moisture will run off without freezing, since the frost layer would injure the cooling efficiency. Under these conditions the gas contains 0.7% moisture. For lower moisture contents activated alumina is used, Fig. 27, since it will dry gas as thoroughly as though it had been chilled at around $-40°$, i.e., to around 0.01% H_2O.

Sulphur. Contamination of the protective gas by SO_2 leads to scaling or staining, and S-free fuel gases must be chosen, or elaborate precautions taken to remove even traces of H_2S, mercaptans, or other S-containing constituents from the gas. City gas whose S content is satisfactory for its use as household fuel may be much too high in S to be used as a source of a controlled atmosphere.

Summary. *Bright Annealing.* Bright annealing of stainless steel or any steel high in Cr is a special case, since these steels are stained by CO. The protective gas should be pure H_2 or a $H_2 : N_2$ mixture. The H_2 must be very carefully freed from any trace of free O_2 and must be thoroughly dry,[47] e.g., dried over activated alumina. (Low-Cr steels[25] will also be stained in gases containing CO, but this is not necessarily accompanied by decarburization, so from the point of view

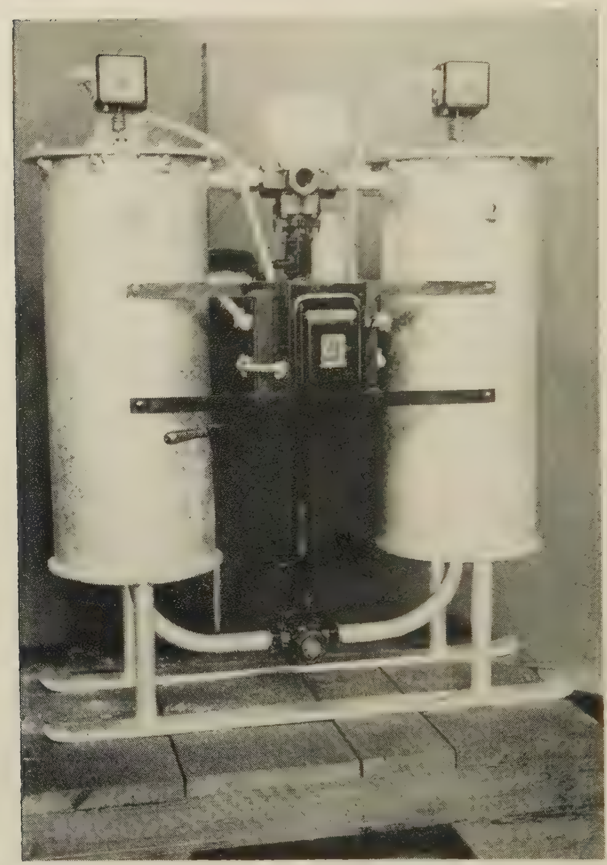


FIG. 27.—Activated alumina dryer. Two chambers are filled with activated Al_2O_3 and provided with heating elements. When one chamber becomes so saturated with moisture that it no longer dries the gas properly, the gas is switched to the other chamber and the heating current turned on in the first chamber, to drive out the moisture and prepare the Al_2O_3 for further service in its turn. (Courtesy Pittsburgh Lectrodryer Co.)

of hardening without decarburization, the cheaper CO-containing gases are serviceable for the ordinary low-Cr steels.)

Bright annealing of low-C steels at 1200–1250° can be accomplished in a variety of atmospheres. The cheap generator gas from partial combustion of a fuel gas with a deficiency of air serves, if it is correctly dried and does not contain any free O_2, or more than half as much CO_2 as CO, since the H_2O-forming reaction of H_2 or CH_4 with CO_2 is not active at this temperature. A higher CO : CO_2 ratio is better to aim at, since fluctuations of gas pressure to the generator or slight changes in the generator temperature may alter the ratio, and it is better to start with some leeway. Troubles with bright annealing are generally due to incorrect drying, or to too high CO_2 in the entrance gas, or to infiltration of air into the furnace. If the inlet and exit gases are not only analyzed but *also* checked for moisture content, which is simple with the dew point meter, the source of trouble should be easily traced. Operators are often inexcusably ignorant of the composition of the exit gas.

A small amount of water vapor, that in equilibrium with the gas at the temperature used, is desirable. If that tiny amount is not present, the reactions $2\ CO \rightarrow CO_2 + C$, and $H_2 + CO_2 \rightarrow CO + H_2O$ will be catalysed at the surface of the steel and, whereas no staining may occur, the steel will become slightly dulled or etched to a matte surface, as Marshall [27] and Otis [48] have pointed out. This need for a trace of moisture at the annealing temperature is opposed by the necessity of avoiding an excess of moisture during the cooling period, so that a very close balance is necessary as to H_2O content of the gas.

Bright Normalizing. Bright normalizing of low-C steels at 1700–1750° requires that the gas shall be initially dry and also sufficiently low in CO_2 so that H_2O, beyond that which can be balanced by the H_2 present, will not be formed by reaction of CO_2 with H_2 or CH_4. The tolerance for H_2O depends on the rate of cooling of the steel, since the staining occurs during cooling and the steel may come out clean if cooled too fast for the reaction to produce a visible stain, even in an atmosphere that would stain on slow cooling. Removal of CO_2 from the gas would make bright normalizing more fool proof.

Spheroidization. A long spheroidizing anneal of medium- or high-C steel in which decarburization is to be avoided is conveniently carried out in a suitably balanced $H_2 : CH_4 : N_2$ that is thoroughly dry. Exact information as to the behavior of generator gas in this service is lacking. Such gas must certainly be thoroughly dried, but the tolerance for CO_2 at these times and temperatures is not well established for gases containing H_2 as well as CO. Since carburization goes on so slowly at these temperatures, gases of the strongly carburizing type would be relatively inert. It would appear safer to remove CO_2 from generator gas for this service.

Hardening without Decarburization. Heating medium- and high-C steels for hardening without decarburization requires careful choice of gas, the choice being the more limited the higher the C. Thorough drying of any gas is essential; moisture must be kept out. In order to remove the opportunity for H_2O formation in the furnace itself, CO_2 must likewise be kept out, or held to a low value in relation to the CO, in order to prevent the reaction $CO_2 + H_2 \rightarrow CO + H_2O$. Furnace makers[49] are beginning to realize that scrubbing out CO_2 from generator gas is as imperative as drying the gas, and CO_2-scrubbing equipment has had to be provided in installations not originally supplied with it. Severe decarburization has thus been overcome, so there is ample commercial, as well as theoretical, evidence for the necessity of purifying the gas from CO_2.

The degree of purification necessary varies with the C content of the steel to be treated and the CO content of the protective gas. Charcoal gas, with its higher CO, does not have to be as completely freed from CO_2 as does generator gas.

Compensation for the decarburizing tendency of $H_2 : N_2$ or $CO : H_2 : N_2$ can be accomplished by small additions of CH_4. The presence of much CO_2, however, prevents compensation.

High-speed Steel. For the high-speed steel preheat, gases suitable for hardening ordinary steel without decarburization are applicable, but are as yet little used. Slightly carburizing atmospheres for the high heat seem theoretically advisable but the short time at heat ordinarily requires a batch rather than a continuous, conveyor type furnace, and it is difficult to avoid infiltration of air in charging and discharging. Hence maintenance of sufficiently oxidizing conditions to avoid decarburization is generally adopted and the formation of some scale accepted. Further experience in hardening ordinary steels without either scaling or decarburization will doubtless lead to methods for similarly handling high-speed steels.

Solid Packings. In small-scale operation, fair protection against decarburization may be secured by packing the work in cast-iron chips. Packing in charcoal, spent carburizer, or active carburizer may also be utilized, with proper regard to temperature of operation and C content of the steel in selecting the packing.

Contamination of Gases. Sulphur-bearing gases should be avoided in choosing a gas for a controlled atmosphere. Moisture or oil on the steel treated will affect the atmosphere and should be removed, or, if feasible, the atmosphere adjusted to compensate. Scale or rust on the steel treated, under usual operating conditions, will be reduced in the protective atmospheres, its O_2 forming H_2O or CO_2. For example, a steel previously normalized in air and scaled but not decarburized, heated for hardening in a controlled atmosphere that would not decarburize clean steel, might produce enough H_2O to cause serious decarburization. Steel should be clean before trying to handle it in a controlled atmosphere. Some investigators think that gases adsorbed on the surface of the steel, or even evolved from the body of the steel, play a part, and need to be taken into account in special cases.

Heat-resistant Alloys. A gas chosen to protect the steel may be entirely suitable for that purpose, but may have a deleterious effect on resistors, radiant tubes, conveyors, or other furnace parts made of heat-resistant alloys. This problem is not well enough understood at present to allow statements as to what limitations must be set in order to secure the longest life of the furnace combined with satisfactory protection of the work.

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CHAPTER 13

CYANIDING, NITRIDING, SURFACE HARDENING, COATING

Fused Baths. One way to avoid the effect of air or flue gas atmospheres is to prevent their presence by heating the work by immersion in a fused bath.

Molten lead can be used as a heating medium from about 650° to perhaps 1700°. In rare cases it is used at even higher temperatures. The higher temperatures involve danger of lead poisoning from lead or lead oxide fumes. Since lead is heavier than steel, the steel has to be held beneath the lead by some means, as by wiring it to a rod and clamping the rod in position. Since there is no "atmosphere" present, and clean, pure lead does not alloy with or attack steel, there is neither scaling nor decarburization within the bath. In order to prevent oxidation of the lead, the bath is covered with charcoal or similar reducing agent. Dirty lead will both cling to the work and decarburize it. At high operating temperatures the work may oxidize in the air during transfer to the quenching bath, or, of course, on air cooling, so that, to avoid all scale or tarnish, transfer or cooling might be done in a controlled atmosphere.

Rate of Heating. Beside the freedom from atmospheric attack, lead and other fused baths have the advantage of much more rapid heating than is the case when the same object is heated by radiation. In radiation heating, the side of the object that "looks away" from the source of heat, or a part that is shaded from direct radiation, has to be heated by conduction from the more highly heated portions of the object or by slow conduction from the furnace atmosphere. In the lead bath all sides of the object are heated simultaneously and, since the thermal conductivity of lead is high, the bath may be kept at the temperature to which the steel is to be heated and sufficient heat transmitted through the lead to maintain it at that temperature without being strongly cooled locally by the insertion of the cold steel. (See Fig. 1.)

The uniformity of heating in lead allows irregular objects to be heated more rapidly without cracking or without overheating thin edges than is the case in an ordinary furnace.

Fusible Alloys. Lower temperatures than 650° can be had in fused metal baths of lead and tin, or in the very low melting "fusible alloy" baths. The lead-tin baths, approaching the terne plate coating composition or even the composition of solder, are much more likely to "wet" the steel and adhere to it than pure lead. The lower the temperature of the lead bath, the less fluid it is and the more likely to show some adhesion to the metal. Dipping the steel in brine and heating so as to cover it with a film of salt as the water evaporates, or coating with some wash to prevent adhesion, has been practised but is a dangerous procedure since the steel must be absolutely dry when immersed in the bath to avoid spattering. Adhering lead may be removed by dipping in fused caustic soda (NaOH).

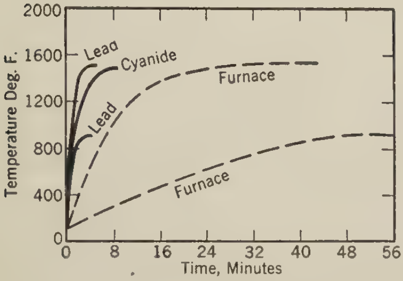

FIG. 1.—Relative rates of heating of steel in lead, in cyanide, and in an ordinary furnace. (After A.S.M. Handbook.)

Lorig[1] has suggested the use of molten cast iron as a high-temperature "pickle" for removing scale or dissolving away a decarburized skin, e.g., on billets for rolling or bars for forging.

Salt Baths. Fused salts can be made up to cover a wide range of temperatures, from as low as 300° for tempering to as high as 2350° for hardening high-speed steel.

Those for the lower temperatures are made up of sodium and potassium nitrates ($NaNO_3$-KNO_3) and for the very lowest temperatures sodium nitrite ($NaNO_2$) is added in varying proportions. These are strongly oxidizing at high temperatures and are not used much above 1000°.* Calcium and barium chloride mixtures ($CaCl_2$—$BaCl_2$), or straight $BaCl_2$, may be used at high-speed hardening temperatures, but the pot life is short and externally heated pots for salt bath hardening for high speed have found little use. Refractory lined containers carrying salt baths that are internally heated by their own electrical resistance have been used, but the service is hard on the refractory linings. Very satisfactory behavior of salt baths electrically heated so as to produce circulation is reported by Hultgren.[4] A borax film produced by dipping the steel into a saturated solution, or by sprinkling powdered borax over it, is resorted to in special cases, such as in the forging, and hardening of molybdenum high-speed steel.[5, 6] The borax fuses and forms a protective coating over the steel.

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* Schneider and Linden[2] and Mailänder and Ruttmann[3] report that when steel is left in nitrate, or nitrite-nitrate baths for long periods, as in former German practice in creep testing, considerable amounts of N may be taken up.

In the more commonly employed baths for operation at 1100–1650° the chief constituents are sodium carbonate, sodium chloride, and calcium or barium chloride (Na_2CO_3, NaCl, $CaCl_2$—$BaCl_2$). A bath of ⅓NaCl—⅔ Na_2CO_3 may be used in the range 1200-1500°, one of ½ $CaCl_2$—½ $BaCl_2$ for 1200-1650°, one of ⅓ NaCl—⅓ $NaCl_2$—⅓ $BaCl_2$ for 1100-1650°. Fused sodium hydroxide (NaOH), melting at about 600°, is used for cleaning steel wire prior to zinc plating, in the Tainton process,[7] and pure Na_2CO_3, melting at about 1560°, has been suggested for combined cleaning and annealing.[8]

For heating C steel for hardening, the NaCl-Na_2CO_3, the $CaCl_2$-NaCl or $CaCl_2$-$BaCl_2$, and the NaCl-$CaCl_2$-$BaCl_2$ mixtures are most used.

The commercial chlorides normally carry small amounts of sulphate, and, since sulphates are corrosive, just as SO_2 or SO_3 is in controlled atmospheres, the sulphate has to be eliminated, as by allowing the bath to act on cast-iron chips until the sulphate is exhausted.

The chloride and chloride-carbonate baths have some tendency to corrode and pit the steel. This tendency is sometimes ascribed to the solution in the bath of oxygen and water vapor from the air. The baths tend to decarburize at 1500° and have quite a strong decarburizing action at 1600°.

The salt baths tend to adhere to the work, so there is some "drag-out," but this adherence protects during transfer to the quenching bath or during cooling. The coating may interfere with quenching, some of the salts are hard to wash off, and the chlorides may cause rusting of the steel. These difficulties have to be contended with or overcome in some way.

Cyanide Baths. To minimize or avoid the decarburizing tendency, small additions of borax or boric oxide (B_2O_3) or of sodium cyanide (NaCN) may be made from time to time. NaCN, on decomposition in a melt open to the air, changes to Na_2CO_3 and, since more NaCN has to be added from time to time as the cyanide becomes depleted, the bath builds up in Na_2CO_3. Thus one might as well start with the NaCN-Na_2CO_3 bath; and this, plus slight NaCN additions from time to time, is chosen when a neutral bath in which avoidance of decarburization, together with the slightest possible carburization and nitriding, is sought. However, NaCN is more often used in larger proportion, in order to secure actual case hardening.

Another way to produce a neutral salt bath, for heating, for quenching, or for annealing without scaling or decarburization, is to use one of the chloride mixtures and periodically add calcium cyanide $Ca(CN)_2$. These are called "activated" baths.

Thus, it is possible, by making small periodic additions of cyanide, to combat the decarburizing tendency of a salt bath and produce "clean hardening." The point of neutrality is generally slightly overstepped but, with the short heating period used, only a trace of case is formed. The wide use of such processes indicates that, in controlled atmosphere heating, the use of atmospheres that are slightly on the carburizing side should also be quite generally applicable.

Quenching from Cyanide Baths. Carburized parts that are not quenched direct from the carburizing temperature may be reheated in a cyanide or activated bath for quenching, in order to prevent scaling or decarburization. McQuaid and McMullen[9] state that the surface hardness of carburized steel is decreased rather than increased by such a cyanide treatment, as shown by tests on carburized 1020 and 4615. Portevin[10] finds that cyanide baths can carburize low-C and decarburize high-C steels. Bishop[11] reports the same thing.

With equal depths of case, the N-containing cases produced by the cyanides are more brittle than the usual carburized cases. The C-N cases are therefore seldom produced of great thickness, and are applied chiefly to medium-C, rather than to low-C steels. The medium-C steel core is heat treated to produce high strength and hardness rather than great toughness, the strength and hardness of the core being required to support the thin case.

Cyanide Cases. Since in many applications like automotive gears, for example, very slight wear ruins the object, the high-C, wear-resistant skin need be only very thin. In such cases it is necessary to carburize only for a short time, but to be sure that in that short time a surface skin of the proper C or C and N content is built up. Solid carburizers, as was shown in Chap. 10, do not readily build up a hypereutectoid case in a short time, so for very thin wear-resistant skins, either gas carburizing which can be adjusted to produce a hypereutectoid case, or a cyaniding process is chosen. Cyaniding in baths of *low* cyanide content does not build up a thick case as rapidly as do the box and hydrocarbon gas methods.

Schenck[12] prefers a cyanide case of about 0.002 in. on gears, on the theory that the least thickness that will provide wear resistance is the best. This is built up on S.A.E. 3145 steel by heating in cyanide 6 min. at 1325°, transferring in turn to two other cyanide pots, both at 1500°, and finally to another at 1325°, after which the gears are oil quenched. The NaCN content used is low, being held between 3 and 6%. The depth of case increases with time, temperature, and with concentration of cyanide in the bath. Thus 30 min. at 1500° will give 0.002 in. in 5% cyanide, 0.004 in. in 10%, and 0.006 in. in 30%.

Using 10% cyanide, a 30-min. dip at 1300° will give 0.0005 in., at 1400°, 0.0025 in., at 1500°, 0.0040 in., and at 1600°, 0.0048 in. Stronger cyanide and shorter times are commonly used. Cyaniding is usually done just above the upper critical for the core and the work quenched directly. The C in the case will run from 0.50 to 0.75%, but about 0.50% N will also be taken up. The case is file hard because of the presence of iron nitride in spite of the low C content.

Cyanide Case Depth. According to the "A. S. M. Handbook," case depth in cyaniding does not vary greatly with the composition of the steel. S.A.E. 1020, 2315, and 3115 all take on approximately the same depth of case under the same conditions, though the Ni and the NiCr steels may show slightly lower C in the case and a better gradient. The cyaniding case depths given by the "A. S. M. Handbook" for different times and temperatures are:

TEMPERATURE Deg.	TIME Hr.	CASE DEPTH In.	CARBON CONTENT OF CASE %
1550	1	0.10 –0.012	0.60
1550	3	0.018–0.020	0.70
1600	1	0.010	0.45
1600	3	0.015–0.018	0.65

Cyanide treatment is advocated by Gill [13] for fine-edged, high-speed steel tools like taps and reamers. The finished tool is slowly preheated to 1000° and immersed for from 5 to 60 min. in an aged bath, originally made up of 50% NaCN and 50% KCN, held at 1050°. The tool is then cooled in air. The case is so thin that the method is applicable only to finished tools.

NaCN appears to decompose on heating in two different ways, according to Barsky,[14] though both are brought about by the effect of oxygen from the air.

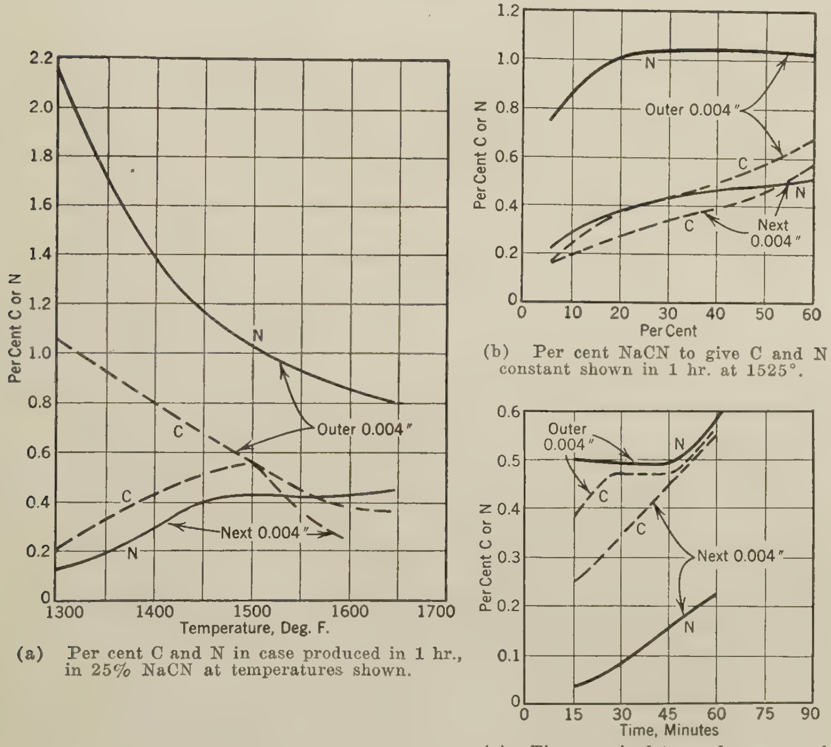
$$(1)\quad 2NaCN + 2O_2 \rightarrow Na_2CO_3 + CO + 2N$$

$$(2)\quad 2NaCN + O_2 \rightarrow NaCNO;\ 2NaCNO \rightarrow NaCN + Na_2CO_3 + C$$

The effect of oxygen was checked by noting that in a deep bath action was less rapid at the bottom, where air had less access, and that action was slowed down if the air over the pot was displaced by N_2.

Composition of the Case. Nascent N and nascent CO and C appear to be produced at about the same rate, since the case contains approximately equal parts of each.

Bishop [11] has traced the effect of temperature, time, and concentration of NaCN on the relative amount of C and N in the case. His data are shown in Figs. 2*a*, *b*, *c*. Nitriding by nascent N is a relatively low-temperature operation, while rapid carburizing requires high temperatures. Cyaniding temperatures are in between, and the ratio of N absorbed to C absorbed will be higher the lower the cyaniding temperature. As the cyaniding temperature is increased the C absorption becomes the preponderant action.



(a) Per cent C and N in case produced in 1 hr., in 25% NaCN at temperatures shown.

(b) Per cent NaCN to give C and N constant shown in 1 hr. at 1525°.

(c) Time required to produce case, of C and N content shown, in 36% NaCN at 1575°.

FIG. 2.—Carbon and nitrogen in outer 0.004 and next 0.004 in, in cyanide cases produced on S.A.E. 1020 steel under varying conditions. (After Bishop.[9])

The case depth produced in a 50% NaCN, 50% Na_2CO_3 bath, with varying times and temperatures, as given by Thews and Snelling,[15] is shown in Fig. 3. Thews and Snelling state that 3% Ni steel is more rapidly cased by cyanide than are plain C steels. If this is correct, it is just the reverse from ordinary box carburizing.

Commercial Baths. A widely used commercial liquid carburizing bath of the type used for heavy cases is reported to give more rapid rates of penetration than those shown in Fig. 3, the depth curves for 1500° and 1650°, respectively, being about the same as those shown in Fig. 3 for 1550° and 1750°.

The penetration at 1600° in this liquid carburizer is a bit over 0.01 in. in the first half hour, 0.015 in. in 1 hr., 0.031 in. in 3 hr. When a case of 0.03 in. is produced, the outer 0.004 in. contains about 1.1 C, 0.20% N; the second, 0.90 C, 0.15% N; the third, 0.80 C and 0.10% N.

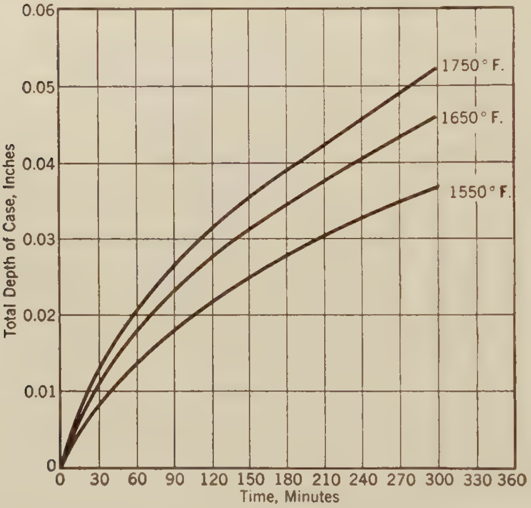


Fig. 3.—Depth of case vs. time in 50 NaCN, 50 Na$_2$CO$_3$. (After Thews and Snelling.[15])

Another commercial carburizing bath, sold on a trade-name basis and not described in scientific publications, is claimed to carburize only to the eutectoid or pearlite composition and to give a "true iron-carbon case" that is free from nitride needles, hence very tough. Without stating that N does not enter at all or that cyanide is not used, the advertising literature is so phrased as to lead the reader to think that the cyanide reaction is not involved and that the specially treated C, added to float on the bath and protect it from oxidation, is the active carburizing agent. Analysis shows that the bath does contain cyanide, and additions to make up for drag-out will maintain a moderate cyanide content. Representatives of the firm marketing the material admit that the bath is not cyanide free and that the case may contain some N.

Claims are made in the patent literature relating to this product to the effect that a phenomenon occurs similar to that in box carburizing, i.e., transfer of C from solid C, added in pellets that float on the bath, to the steel, through gas reactions. It is extremely unlikely that such a phenomenon can be effectively brought into play. The mechanism involved is probably the usual one of decomposition of cyanide.

The makers of this carburizing bath recommend it for rapid carburizing up to 0.04 in. case depth. They state that on steels of 0.12-0.20% C cases of approximately the following depths may be obtained in the times stated at 1600°.

Time	Inches, Case Depth
20 min.	0.005–0.010
40 min.	0.010–0.015
1 hr.	0.015–0.020
2 hr.	0.020–0.025

The method of measuring case depth is not stated. Houdremont[16] presents data to show that, at the same temperatures and times, such a bath produces as thick a case as does a strongly energized solid carburizer.

This bath is reported by users to give satisfactorily tough cases. Hence this example and the analysis of the case given by the other commercial bath referred to above indicate that, by proper choice of bath, cyanide concentration, and operating conditions, the amount of N taken up can be held to a lower value than is ordinarily met in cyanide hardening. It cannot be taken to indicate that the solid-carburizer reactions can be made to proceed in a fused bath without simultaneously calling upon the usual cyanide reaction.

Activated Baths. The "activated" baths, chloride salt baths, which use calcium cyanide $Ca(CN_2)$ instead of NaCN react primarily according to the following reaction: $Ca(CN_2) \rightarrow CaCN_2 + C$, though some formation of nascent N occurs also. However, the liberation of nascent C is greater than the liberation of N, and the surface cases contain, depending on time and temperature, around 0.70 to 1.0% C, together with about 0.20-0.25% N.

Thus the cases from activated baths are less brittle than those from the NaCN baths and approach more closely to those obtained in box or gas carburizing.

Case Depths. The cases from the $Ca(CN)_2$-activated baths grow to a given depth somewhat faster than those from NaCN, the case being some 25 to 50% deeper for the same time and temperature. That is, C diffuses in faster than N, again showing that the activated bath behaves more like ordinary carburizing than does the NaCN bath. Because of this difference in diffusion rate, the C-N cases tend to be richer in N on the outside.

Beckwith [17] shows Fig. 4 for the relative rates of casing in cyanide and activated baths. He recommends 1450-1550° for operating cyanide baths, stating that the automotive industry holds to that temperature range. In discussion, Tour [18] argued for a lower temperature range, 1350-1450°, because of rapid decomposition of cyanide.

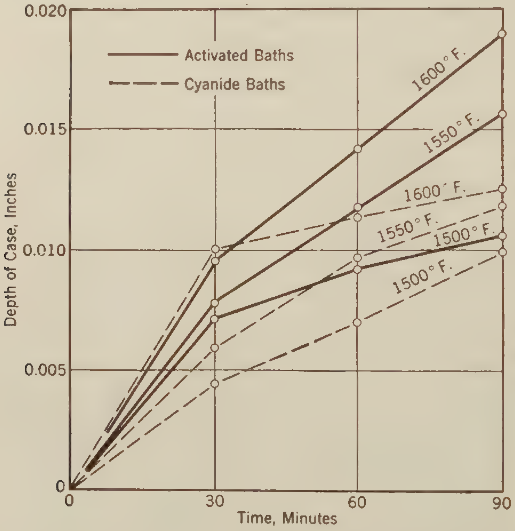


Fig. 4.—Data obtained on $\frac{9}{16}$ in. bars in commercial-scale tests. The cyanide bath contained 22–24% NaCN. Depth is average for S.A.E. 1015, 3115, and 4615 steels. Hardness, determined by Rockwell Superficial Tester (15N scale), was the same (89.5–92.5) for all steels, all times of immersion, and for either bath. (Beckwith.[17])

A crude form of cyanide hardening, used by blacksmiths to produce a very thin case, is to heat the steel to 1550-1600°, plunge it into powdered NaCN (96% grade) or sodium ferrocyanide, or sprinkle one of these compounds on the hot steel, reheat to 1550-1600° to melt the compound, repeat if necessary, and quench.

Cyanide Is Poisonous. The poisonous nature of all cyanides must be remembered and the greatest care taken not to get them into the mouth or into cuts or sores. The furnaces should be hooded and ventilated so that fresh cyanide fumes are not breathed. The steel heated in a cyanide bath may be quenched into limewater to neutralize the adhering cyanide and prevent danger in handling the steel thereafter.

Good cyanide cases are file hard, but not superhard. For cases of extreme hardness, the nitriding of alloy steels is resorted to.

NITRIDING

We have seen that, in cyanide and activated baths, nascent N forms hard cases on steel, that such cases tend to be brittle, that the diffusion rate of nitride is low, so that nitride-containing cases are ordinarily kept thin, and that a harder core is generally necessary to give proper support to a thin, brittle case than is necessary with a heavier, well-graduated C case.

Hardness of Nitrided Case. An extraordinarily hard case of as high as 1100-1200 Vickers-Brinell, and one that is extraordinarily wear-resistant, can be produced by the action of nascent N only. The hardness, compared to a case carburized steel is shown by Fig. 5*a* and that for two different steels in Fig. 5*b*. The nascent N is obtained by the thermal decomposition of anhydrous ammonia, NH_3. The NH_3 gas is led into the closed, externally heated container in a manner analogous to that used in gas carburizing, and dissociates in contact with the work. The temperature used is very much lower than the

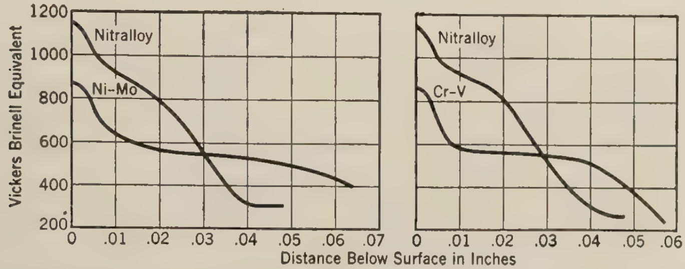


Fig. 5a.—Hardness of nitrided vs. case carburized steels. Nitralloy, heat treated, then nitrided 90 hr. at 975°. NiMo, carburized, double oil-quenched, drawn at 300°. CrV, carburized, double oil-quenched, drawn at 300°. (Grossmann.[19])

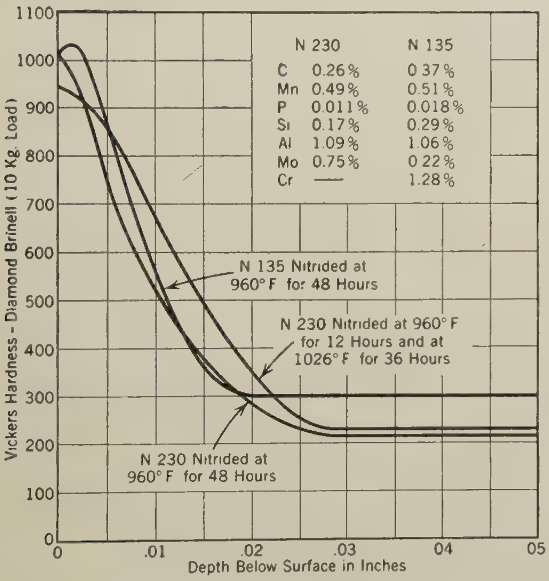


Fig. 5b.—Hardness of nitrided steels.

temperatures used in carburizing and cyaniding. Steel is ferritic, not austenitic, during normal nitriding.

Use of Alloy Steels. Plain C steel may be given a nitride case, but the case is too brittle to be useful. Steels for practical nitriding always contain one or more of the elements that form nitrides very readily, Al, Cr, or V. Nitriding is carried on in the range of temperatures at which temper-brittle steels develop their temper brittleness. Mo, as will be shown in Chap. 9, Vol. II, is a specific against temper brittleness. For this reason in respect to the core and because the presence of Mo is almost equally a specific against excessive brittleness of the nitrided case, about 0.20% Mo is included in the common Al-Cr nitriding steels. Or, Cr can be omitted, and the Mo raised in an Al-Mo steel, or the Al can be omitted and Cr plus V used. Si, Mn, Ni, and Cu need not be excluded, but they do not facilitate nitriding. The total content of Al, Cr, Mo, and V usually sums up to from 2 to 3%, though very much more highly alloyed steels may be used. The C content ordinarily runs from 0.20 to 0.40%, which in the presence of the rather large amount of alloying elements makes the steels readily heat treatable, and allows tempering temperatures high enough so that the core will not soften during nitriding too much to fail to support the case. By adding some Ni to an Al-containing steel, precipitation hardening, due to Ni-Al compounds, may be secured at nitriding temperature, to afford added strength to the core (see Chap. 8). High S may be present to make the steel free machining, without interfering with nitriding. High-speed steel may be nitrided, and Gill [13] suggests such treatment for tools to be used for cutting bakelite. For most purposes the nitrided edge is too brittle, so cyaniding is generally preferred.

Preparation for Nitriding. The presence of ferrite, free from emulsified carbide at the surface, prevents the formation of a satisfactorily tough nitrided case. The best type of case results from having the steel, before nitriding, in the sorbitic condition, after quenching and drawing. The usual nitriding steels may be forged at 1950-2200°, quenched from proper hardening temperature, and drawn at 1150-1350°. If during heating for quenching, the steel has been given a decarburized surface, that ferrite surface must be removed before nitriding. Use of suitable controlled atmospheres in the heating for quenching is therefore indicated.

Surface Defects. If the decarburized surface is not removed before nitriding, the case will spall very readily. This condition may arise because of improper centering during the machining whereby an insufficient amount of material is removed from some points. This may occur in the machining of forgings. Parts that are coated with rolling or forging scale necessitate the removal of $\frac{1}{32}$ to $\frac{1}{16}$ in. of material to eliminate the decarburized layer.

Although decarburization prior to nitriding is fatal, the nitriding steels, such as the common Al-Cr-Mo composition, are readily decarburized in the nitriding process. The NH_3 in decomposing to nascent N and nascent H naturally has a decarburizing action, such as Chap. 12 has brought out for H_2 in general; though at the relatively low temperatures of nitriding the effect is not great. McQuaid and Ketcham[20] note that a 0.33% C nitriding steel with Al and Mo lost half of the C from its surface in 20 hr. at 950°. Carbon seems to be removed from the surface during nitriding, simultaneously with the addition of N. This fact may explain the phenomenon often noted, that the extreme surface of a nitrided case may not be quite as hard as the layer next beneath, as if, at the zone of maximum hardness the hardening effect of both the C and the N were reinforcing each other.

Carburizing-nitriding Procedures. Machlet[21] has suggested adding hydrocarbons to oppose the decarburizing tendency of the hydrogen freed from the NH_3, but as we know from our discussion of gas and pack carburizing, the carburizing action at 950-1000° is negligible, while decarburization can occur at low temperatures.

Berkhin and Laklitin[22] claim that bubbling the NH_3 gas through aniline ($C_6H_5NH_2$) before it enters the nitriding chamber accelerates nitriding. Whether any carburization occurs in such a process is not stated.

Egan[23] has reported the production of a case 0.007 in. thick, with 1000 Vickers-Brinell surface hardness, on a 0.25 C, 1.69 Cr, 0.63% V steel nitrided by NH_3 bubbled through $C_6H_5NH_2$, by only 4 hours' exposure at 1580°.

Straight aniline vapor, from aniline fed into an electrically heated gas carburizing outfit, was found, in tests kindly reported by H. E. Koch, to produce a case about 0.0025 in. thick in 4½ hr. at 1500°. The case, quenched as for regular case hardening, showed 900 Vickers-Brinell at the very outside, and again at 0.0015-0.0020 in. depth, but from 0.0002 to 0.0014 in. depth the hardness was 850. The distribution of C and N in the case is not known, but the hardness curve would indicate that both elements were taken up.

The possibility of simultaneous gas carburizing and nitriding by leading in both a hydrocarbon gas and NH_3 is being examined by at least two groups of furnace makers. According to the few discussions that have been published,[24, 25] a thin case may be produced at 1150°, while heavier ones are produced at 1250-1500°. For some applications a sufficiently hard case may be secured without quenching. Wyzalek [26] states that gas carburizing with propane and air, to which NH_3 vapor is added, gives a case very similar to that obtained by cyaniding. On S.A.E. 1010, file-hard cases of 0.001 to 0.01 in. are produced without resorting to quenching.

Liquid-bath Cyaniding and Nitriding Combined. A process called "Chapmanizing" that has been very little described outside trade promotion literature consisted, in its early development, in bubbling NH_3 into a cyaniding salt bath, in order to secure more N in the case and hence a harder case than in ordinary cyaniding, and to produce the case more rapidly than in ordinary nitriding. Securing suitable contact of the stream of NH_3 gas with the work immersed in the salt bath is not easy, and users commented upon irregular cases in the early stages of development of the process. Some users consider the process as at present worked out quite satisfactory for thin cases.

Figure 6 shows curves as given by Rolf.[27] A surface hardness of 890 Vickers-Brinell upon a 0.20% C steel casting is reported by Kerr [28] in studies of erosion by cavitation, which is in agreement with the curves shown.

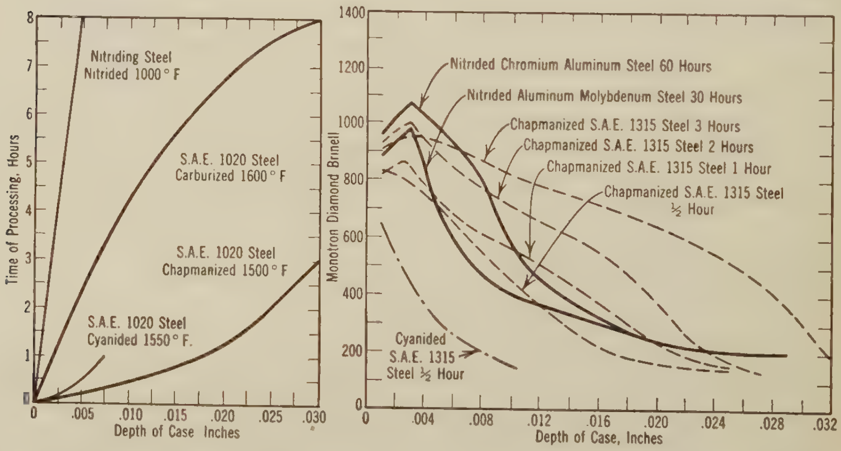


FIG. 6.—Relative rates of producing cases and relative hardness, in cyaniding, nitriding, and "Chapmanizing." (After Rolf.[27])

According to Rolf [27] it would appear that the NH_3 is now dissociated by heating before the gas is led into the bath, the tubing being made of some material that will not deactivate the active N too rapidly. Rolf states that light cases consisting almost entirely of nitride rather than carbide may be obtained at 1250-1400°, but the usual case is best produced at 1500°. The rate of formation of the case upon a 0.17% C steel without special alloys is given as follows:

DEPTH OF CASE IN INCHES

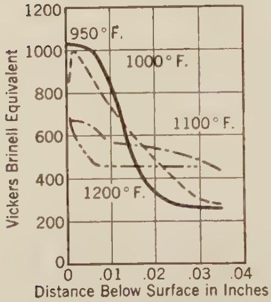
Time, Min.	1450°	1550°
30	.005	.008
60	.011	.017
90	.016	.021
120	.019	.026
150	.023	.027
180	.026	.031

Roshong[29] reports that introduction of NH_3 into the cyaniding bath gives better wear resistance than results from cyaniding alone and suggests dropping the temperature from 1550 to 1300° before quenching.

Little information from sources that are entirely free from possible bias has yet been published upon the various proprietary processes for producing cases intermediate between straight carburized and straight nitrided cases by modification of the nitriding process, but the urge to approximate the hardness of a nitrided case in a shorter period is so great that some mention is deserved of the attempts to devise a satisfactory compromise.

Re-nitriding not Practised. Returning to straight nitriding, a nitrided part on which close dimensional tolerance is required must be correctly nitrided the first time, or else it is scrap. There is no satisfactory way of "de-nitriding" and then re-nitriding. The case can be de-nitrided to some extent, but on the attempt to re-nitride only a brittle case is produced. The nitrided surface could be removed by grinding or sandblasting and the piece then re-nitrided if the resultant change in dimensions can be tolerated.

Effect of Temperature in Nitriding. Contrary to experience with regular case carburizing, the hardness of nitride cases does not increase with increase in nitriding temperature, as Fig. 7 shows. This figure clearly shows why nitriding temperatures of 950-1000° are normally chosen. However, the cases formed at higher temperature, say 1200°, are both deeper and tougher though less hard. A sojourn at a temperature of 950-1000° to produce a hard case followed by one at 1150-1200° to produce penetration and toughness is advocated by some. Others start at the higher temperature and finish at the lower. Jones[30] advocates 930° followed by higher temperature, up to 1020°. He says that such practice is applicable to the usual steels containing Al but not to those with Cr without Al. Kontorovich[31] advocates a cycle of 930-1150-930° which he claims produces equivalent results in half the time of a continuous 930° treatment. Ordinarily but one temperature is used, that being chosen to produce a case of the desired compromise between hardness and toughness. Lysaght [32] shows Fig. 8.

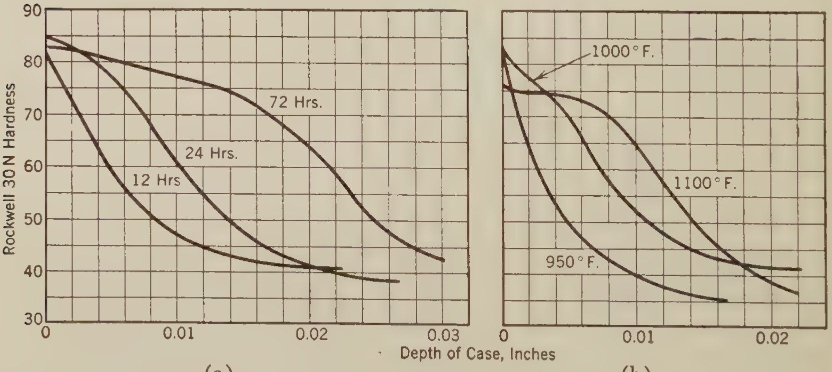


Composition of steel, 0.32 C, 0.63 Mn, 0.15 Si, 1.68 Cr, 0.19 Mo, 0.70% Al.

Temp. deg.	Time hr.	Dissociation per cent	Flow cc./min.
950	48	25	200
1000	48	40	200
1100	48	75	500
1200	48	90	500

Fig. 7.—Effect of variation in nitriding temperature upon hardness. (Grossmann.[19])

Toughness. Toughness of the case is a relative term, since even the "tough" cases, if not well supported, are brittle. When the case is built up locally to too great depth, as occurs at a sharp point or edge, e.g., on a sharp V-shaped thread, the apex will spall readily. Hence parts for nitriding must be made with gradual fillets and avoidance of sharp edges. Nitrided cases are not employed to resist high local pressures unless very thoroughly supported, nor



(a)

(b)

Fig. 8.—(a) Depth hardness of steel nitrided for times shown, at 975°. (b) Depth hardness of steel nitrided 24 hr. at temperatures shown. The steel used was like N 135 in Fig. 5 (b), but contained only 0.25% C. (Lysaght.[32])

when impact or battering service is involved. High-speed tools can be nitrided, as would be expected from their composition, and such a treatment has been suggested for twist drills and the like. The usual criterion of toughness is to note whether the case spalls at the edges of an impression made by the hardness testing of the case, by the diamond cone of the Rockwell superficial hardness tester (30 N scale), the diamond pyramid of the Vickers, or the ¾ mm. diam. diamond ball of the Monotron. The load required to crack the case in the ordinary Brinell test may be taken as a measure of its toughness. The regular Brinell test is of course inapplicable for hardness measurements on so thin a case. A torsion test was considered by French and Homerberg[33] to be an especially searching one by which to study the cracking of the case.

Nascent Nitrogen Required. In the nitriding process it is essential that nascent N in sufficient amount be supplied to the surface of the steel. Ammonia broken down elsewhere in the container supplies only inactive, molecular N_2, so the flow of NH_3 gas must be sufficient to supply fresh NH_3 to break down upon the work at the proper rate. For nitriding at the normal temperature, say 975°, such a flow that the exit gas contains 70% unused, undissociated NH_3, or as it is generally termed, "30% dissociation," is used. In nitriding at higher temperatures the dissociation is greater even though the flow has been materially increased. Too high a dissociation tends to decarburize the case too much. In order to avoid undue decomposition of the NH_3 on the surface of the container, where such decomposition is not wanted, fused silica, vitreous enameled steel, Ni, monel, or other alloys high in Ni are generally used for containers, baskets to hold the work, inlet and outlet tubes, etc.

Time Required for Nitriding. Much effort has been put upon speeding up the rate of nitriding. It is claimed[34] that the presence of Cu gauze or turnings is an accelerator. At best, however, the production of the type of case usually required takes 1 to 2 days. Ashdown[35] secured a case 0.02 in. deep in 25 hr. at 1075° on Al-Mo steel, with a surface hardness of 950 Vickers. In order to get speed of nitriding with depth and toughness of case, a higher temperature than usual was chosen, and maximum hardness sacrificed. This slowing down of production, compared to the hour or so of gas carburizing or the even shorter time of cyaniding to produce cases of comparable depth, together with the increased cost of nitriding steels over carburizing steels, restrict the use of nitriding to those cases where the wear resistance or other properties of the extremely hard nitride cases are worth the delay in production.

Utility of Nitrided Parts. As an example Cone[36] cites the production of nitrided cylinder barrels for airplane engines of a steel of 0.40 C, 1.15 Al, 1.60 Cr, 0.40 Mo, 0.55% Mn, which has been normalized from 1800°, then quenched from 1700°, and drawn at 1250° to put it into shape to nitride, and nitrided 50 hr. at 1000° to produce a case of 800-1000 Vickers hardness. Freedom from warping and the ability to protect the outside of the cylinder against nitriding and thus keep it machinable, together with the great wear resistance of the case, justify its use.

Warpage and Growth. The fact that the nitriding process is carried on below the previous tempering temperature assures freedom from warping due to internal stress, and, since no quenching is required, the production of the nitrided case upon a symmetrical object produces but little warping, if the object has been uniformly heated and the gas flow regulated so that all portions have the same case depth. However, few objects to be nitrided are entirely symmetrical.

There is a slight "growth" during nitriding, and a piece that is not symmetrical, or that is protected against nitriding on one side, may warp in consequence. The growth itself can be compensated for by prior machining to just the proper amount undersize. The growth is about 0.001 in. per 0.01 in. of total case depth, but has to be determined for the example in hand when dimensional changes of this magnitude are of importance. Grinding or lapping the nitrided surface may be resorted to when necessary to correct warpage or to finish to very close dimensional tolerance. Lapping off, say 0.001 or 0.002 in. from the surface, may be desired in order to put the very hardest portion of the case at the wearing surface when a case like that of the 1000° run of Fig. 7 or that shown in Fig. 6 has been produced, or to work down to a layer that is less brittle than the extreme surface. The original unlapped surface is more resistant to corrosion than one with the skin removed. The original surface is quite rust resistant against atmospheric corrosion; it also resists corrosive oils, still, non-aerated, fresh or salt water, and alkalis. Under submerged, aerated, corrosion in fresh or salt water it is less resistant, and it will not resist mineral acids.

Homerberg and Walsted[37] have called attention to the "corner effect" that occurs on nitrided parts possessing sharp corners. The growths on all surfaces combine at the corners to form projections containing a high N content with no supporting core. This portion of the case is brittle and likely to chip. Such a condition should be avoided by providing rounded corners.

Protection against Nitriding. It frequently happens that certain portions of an article are to be protected against nitriding. The most efficacious protective agent is tin or a solder containing 80 parts of lead and 20 parts of tin. The coating should be as thin as possible. Although tin and the alloy of lead and tin melt at a lower temperature than that used in nitriding, ample protection is provided by the thin layer that is held to the article by surface tension. Too much of the protective agent may cause the excess amount to drop onto articles at points where protection is not desired. A liquid flux containing zinc chloride and some ammonium chloride should be used. This flux can be made by adding zinc to a solution containing 500 cc. hydrochloric acid, 250 cc. water, and 15 grams of ammonium chloride until no further action takes place. Parts to be tinned should be entirely free from rust or scale. The procedure recommended in using tin or solder for various parts is as follows:

For protection of the end of a part, clean the part to be protected, apply flux, and dip part into molten tin or solder to the depth to be protected. Remove excess tin or solder with wire brush. For protection of the mid-section of a part, protect ends to be hardened with fireproof varnish (such as Bakelite varnish), apply flux, and dip piece into molten metal to required depth. Remove varnish and excess metal with wire brush. For protection of a very small surface or keyways, apply flux to area, then apply protective agent with a soldering iron.

Nickel plating is an excellent means for providing protection, but the plate must be relatively thick to prevent penetration of N.

Painting with either a mixture of glycerine and tin oxide, shellac and tin oxide, or sodium silicate and chrome ore is also used, although complete protection is not always obtained with these materials. A paste containing lead and tin in a very fine condition in a suitable vehicle gives satisfactory protection.

Stability of Nitrided Case. One outstanding property of nitrided cases is their stability at moderately elevated temperatures. There seems to be no loss in hardness upon heating at least as high as the original temperature at which the steel was nitrided. Homerberg compares this behavior with that of ordinary carburized cases in Fig. 9. The hardness is only slowly reduced on heating even to 1600°.

Other Advantages. This fact, coupled with its corrosion resistance, adapts the nitrided case for faces on valves, for high-pressure, high-temperature steam, and other high-temperature service. Mochel [38] has shown that nitrided steel is extraordinarily resistant to galling and seizure, either against itself or against quite a variety of other alloys.

Although the thin case cannot support a large core that itself has low creep resistance, some of the nitrided steels may be considered for high-temperature service on the basis of fair load-carrying ability, largely conferred by the Mo present.

The very hard case is of course very strong. If we calculate its strength by extrapolation of the Vickers-Brinell, tensile strength relation, it would approximate 500,000 lb./in.^{2} for some of the hardest cases. Even a thin coating of such strong material will notably raise the endurance limit of a nitrided steel shaft in rotary bending, since the maximum stress is at the outside of the shaft. This, plus the resistance to corrosion, makes nitrided steel useful in combating corrosion fatigue. However, if the repeated stress is applied axially, in straight tension and compression rather than in bending, the benefit of the strong but thin surface layer would be small.

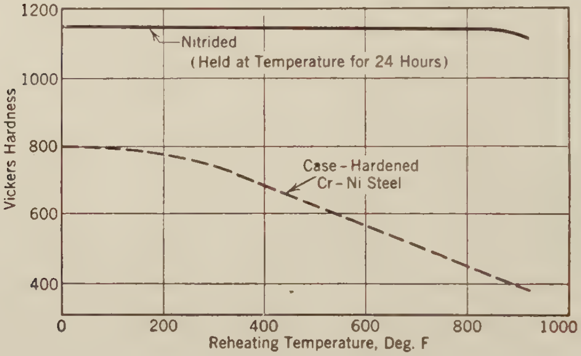
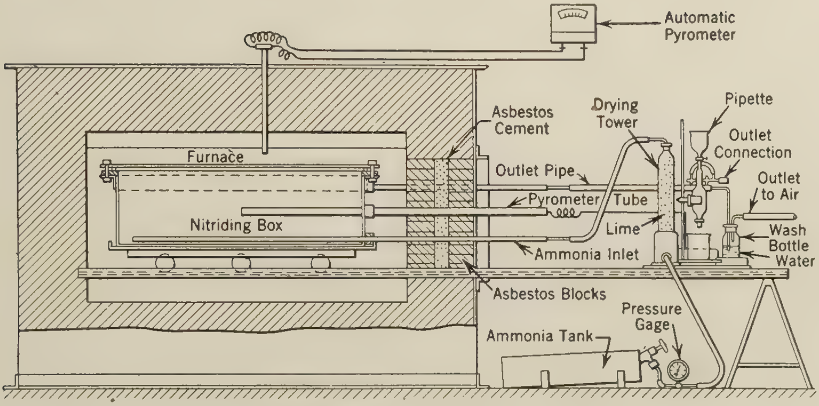


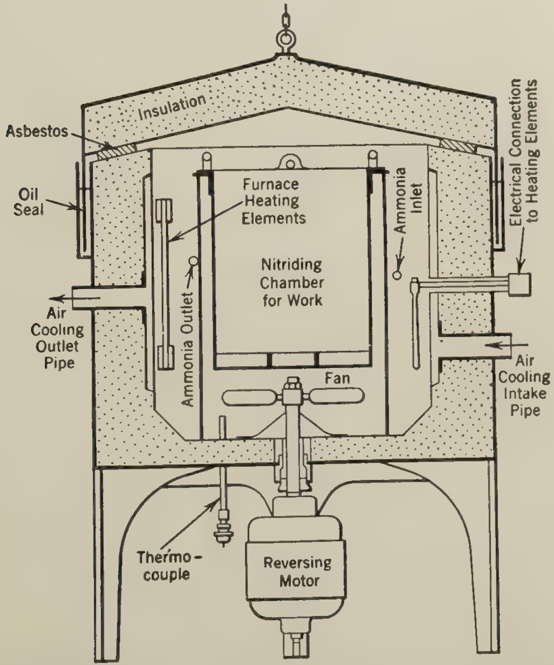
FIG. 9.—Comparison of the hardness of nitrided and carburized cases at room temperature and after reheating as shown for 24 hr.

Nitriding Equipment. The principles of proper heat application, as described elsewhere in this book, apply most emphatically to nitriding. The product to be nitrided must be heated uniformly to and at the required temperature for the stated time in a properly maintained atmosphere capable of forming nascent N.

A typical batch-type nitriding equipment is represented in Fig. 10*a* and *b*. The tank of ammonia must be placed in a position such that the bent tube inside the tank has its open end in the gaseous atmosphere and does not dip into the liquid ammonia. The use of a suitable needle valve ensures the control of a steady and even flow of gas. The inlet and outlet tubes are made of Ni, and the articles in the box are supported on Ni gauze. The inlet tube extends along the bottom of the box to the back; the outlet tube is at the front and near the top of the box. Pyrometers are inserted into both the container and the furnace. The gas mixture from the container passes through a wash bottle to indicate the back pressure, which should be about 1



(a) Diagram of outfit for nitriding.



(b) Usual commercial type of nitriding furnace. See Fig. 25, Chap. 12, for typical nitriding furnaces.

FIG. 10.—Diagrams for nitriding furnaces. (Grossmann.[19])

to 2 in., as well as the rate of flow of the gas, and finally passes into the atmosphere or is washed into a drain by means of a suitable baffle.

Ammonia gas decomposes to a certain extent into N and H according to the following reaction: $2NH_3 \rightleftarrows 2N + 3H_2$. Nitrogen, which is very active at the moment of the decomposition of the gas, combines to a certain extent with the alloying elements in the steel to form nitrides. These nitrides, which are in solid solution or in a fine state of dispersion in the case, impart extreme hardness to the surface of the steel, a hardness which gradually decreases inwardly until it corresponds to that of the core.

The gas mixture leaving the furnace consists of H_2, N_2, and undissociated NH_3. The NH_3 gas is very soluble in water, whereas the H_2 and the N_2 are insoluble. Advantage is taken of this fact in the determination of the extent of the NH_3 dissociation during the nitriding operation. The special pipette shown in Fig. 11 is used for making this determination. The dissociation generally is maintained at approximately 30%.

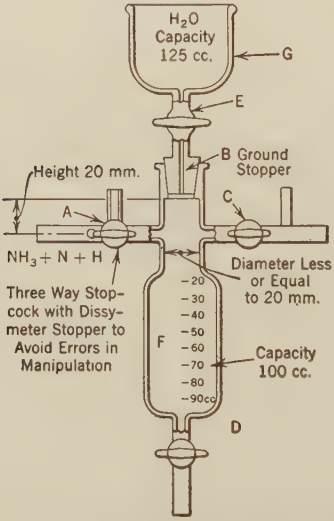


FIG. 11.—Pipette for determining ammonia dissociation.

At the time that a determination of the extent of the NH_3 dissociation is to be made, the stopcocks at *C* and *E* are closed while *D* is open and the stopcock at *A* is turned so as to admit the gas mixture from the container into the graduated chamber. After the air has been expelled, stopcock *D* is closed and *A* is turned so as to by-pass the gas mixture. Reservoir *G* is filled with water, after which the stopcock *E* is opened to admit the water into the graduated chamber. The water immediately absorbs the NH_3, and the atmosphere above this solution contains the insoluble H_2 and N_2. The water solution should reach approximately the 30-cc. mark, indicating 30% dissociation. If it is much above this value the flow should be increased; if much below 30% the degree of dissociation can be increased by decreasing the flow of the gas by means of the needle valve.

Methods of Sealing Containers. Various sealing methods have been discussed by Sergeson and Deal.[39] Figure 12 illustrates these types.

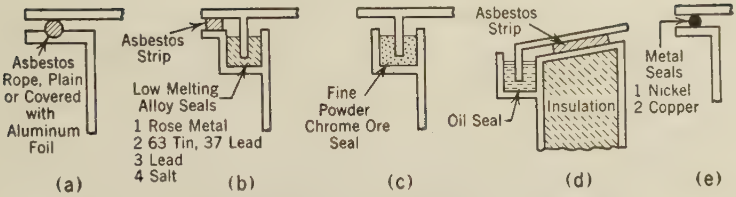


FIG. 12.—Methods of sealing containers.

Precautions and Recommendations. Homerberg sums up the necessary precautions in nitriding as follows:

1. To ensure the best nitriding results, the steel should be heat treated so as to possess a uniformly sorbitic structure.

2. Before nitriding the steel must be free from decarburization.

3. If freedom from distortion is of paramount importance, the internal stresses produced by machining or heat treating should be removed before nitriding by heating to a temperature of 1000 to 1300°.

4. Since some growth takes place on nitriding, this should either be allowed for in the final machining or grinding operation prior to nitriding or removed by lapping or by careful grinding. However, the removal of only a slight amount of the nitride case is permissible.

5. If maximum resistance to corrosion is desired, the parts should be used as they come from the nitriding container.

6. Nitrided steels of the nitralloy type should not be used where resistance to the corrosion of mineral acids is encountered or where resistance to sharp abrasive particles at high velocities is required, as in sand nozzles.

7. If straightening is required after nitriding, it should be done hot at 1200°, although some straightening may be done cold.

8. If maximum hardness and maximum resistance to impact are desired and the question of maximum corrosion resistance is not of vital importance, the removal of 0.001 to 0.002 in. of the nitrided case is desirable. The amount to be removed depends upon the original case depth. This operation will remove the more brittle and somewhat ? softer surface layer.

9. If nitrided articles exhibit a shiny gray surface after their removal from the container, the results should be viewed with suspicion. Invariably, the case will be shallow and below normal hardness. The articles should have a matte gray appearance, although a slight discoloration does not indicate faulty nitriding. A slight vacuum may be created within the container when the furnace is shut off, and, if a small amount of air enters through the gasket, the articles may show various hues such as pink, blue, or light brown. Increasing the flow

of NH_3 at this stage or nitriding in the presence of Cu turnings will frequently prevent this discoloration. The opening of the container at too high a temperature will also cause surface discoloration.

10. Specimens that can be checked for hardness and depth of case should be included with every charge.

11. Applications requiring a high core hardness to support the case when high compressive stresses are present require nitriding steels of special analyses.

12. Although special nitriding furnaces are desirable, such equipment is not absolutely necessary. In most instances these special furnaces are constructed so that they can be used for general heat-treating purposes when not used for nitriding. The most essential point is that a furnace must maintain a uniform temperature within the nitriding container and provide for uniform circulation or flow of the nitriding gases. The container should not react with the gases so as to rob the work of active N, or to produce too much H.

Discussion of steels for nitriding and their properties is given in Chap. 15, Vol. II. The whole topic has been treated in a symposium.[58]

SURFACE HARDENING BY QUENCHING

The widespread use of carburized, cyanided, or nitrided hard skins over a tougher core has naturally led to attempts to produce surface-hardened objects by rapid heating of medium-C steels and quenching them when the skin is above the critical temperature to the desired depth, but the core is below the critical. Obviously, low-C steels cannot be thus treated. Plain C steels of 0.40% C or up are required, though there may be a correspondingly lower C content in alloy steels.

Requirements. In order to avoid too deep hardening it is necessary to establish a steep temperature gradient at the instant of quenching. This means that heating must be very rapid and the quench applied almost immediately after the surface comes to temperature, else thermal conduction will bring too thick a zone above the critical. The flame temperature should be high in order to allow establishing the steep gradient. The oxyacetylene flame is widely used.

Methods Used. Processes in which a gas flame is played on the surface to be hardened and immediately thereafter a water spray is applied are known in this country as "flame hardening" and abroad by the trade names "Shorter," "Double Duro," etc., processes.

Symmetrical objects can be satisfactorily handled; on others warping naturally results unless special precautions are taken. Circular objects, such as piston pins, shafts, and the like are most readily treated. If these can be mounted on centers and rapidly spun under the flame, the whole surface can be heated uniformly, then spray quenched from all sides.

If, however, the object is so large or the shape such that the whole surface to be hardened cannot be rapidly enough heated at the same time, progressive treatment has to be resorted to. For example, large gears may have their tooth surfaces hardened one at a time, with the gear mounted to rotate in a water tank so that the half of the tooth not being heated is immersed to keep it cool. The mounting, burners, and spray-quenching equipment are especially designed for each job and the rate of heating and interval between heating and quenching are mechanically controlled, since these factors cannot be allowed to vary. A weakness of the flame-hardening method when progressive heating, rather than simultaneous heating from all sides, or spinning the object before the flame, is employed is the overlap. On the simplest shape, a cylinder, for example, one can progressively heat and quench as the cylinder rotates, with uniformity of hardening from beginning to end; but, when the complete circuit has been made the heating of the last place tends to soften the first place, so that there is a discontinuity of hardness at the overlap.

Necessity for Mechanization. Whereas many accounts appraise the flame-hardened objects as having a gradual transition zone from case to core and not being subject to spalling, Ricardo,[40] speaking from the point of view of a user automobile engineer, remarks that flame hardening of crankshafts is troublesome to accomplish and very uncertain as to quality. He remarks upon the prevalence of surface cracks that lead to fatigue failure.

Many of the unsatisfactory results met in flame hardening have doubtless come from trying to do the work by hand, or with inadequate mechanization. The time of heating and the exact timing of the sequence of heating and quenching must be most exactly controlled. Mechanical devices for specific applications and some of the precautions necessary for good work are described in recent publications.[41–49] Figure 13 shows burners being mechanically advanced in the heating for hardening of a tooth in a spur gear.

Steels Used. The C content of plain C steels used for surface hardening is ordinarily 0.40–0.60%. In alloy steels containing moderate amounts of depth-hardening alloying elements, the C may be down to 0.35% and, with still larger amounts of such alloys, to 0.30%. Checking during hardening of the more highly alloyed, strongly depth-hardening steels leads to a general preference for the lower alloy steels. Fine-grained steels are usually preferred. S.A.E. 6135, T1340, and

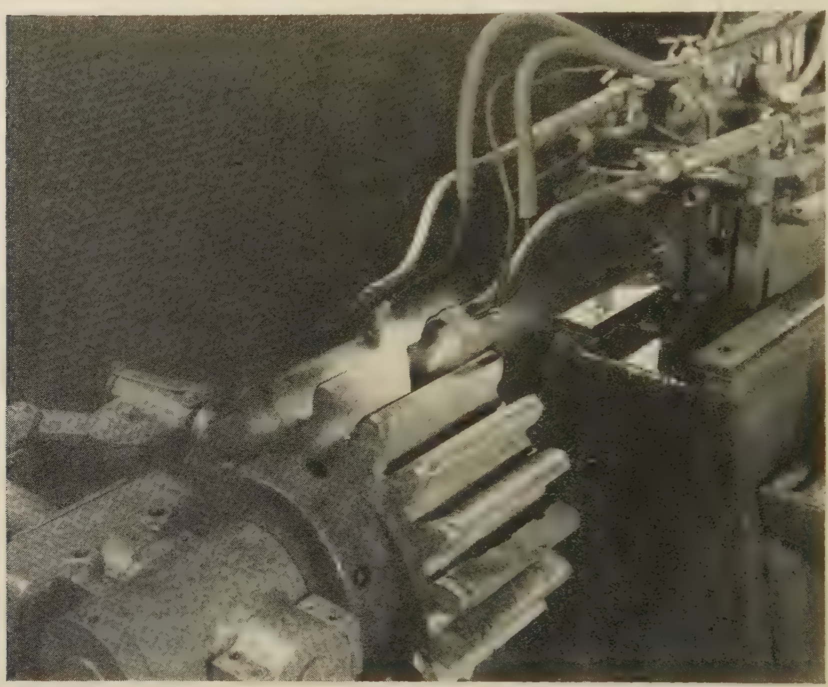


FIG. 13.—Surface hardening a spur gear. (Courtesy A. B. Kinzel.)

T1335 plus 0.20% Mo are representative of the alloy steels. A steel of 0.30 C, 2.50–2.75% Mn has been used on oil-well drill pipe with flame-hardened ends.

A low temperature, stress-relieving draw should be given immediately after hardening. If the piece to be hardened has initial internal strains, care should be taken to relieve them before hardening.

In ordinary practice a depth of 0.1 to 0.2 in. is hardened in the flame-hardening process. Figure 14 shows sections of flame-hardened objects. To get thinner cases, more rapid heating and quenching must be resorted to.

Induction Heating. This can be accomplished by electric induction heating, by which, instead of having to heat by conduction from a flame, heat is generated in the shell of metal to be heated by making it the secondary of a high-frequency induction heating outfit. Equipment for such a process (termed the "Tocco" process) has been developed by the Ajax Electric Company and applied to crankshaft hardening and allied problems, as has been described by Tran and Benninghof [50] and by Cone.[51] The throws of the crankshaft are surrounded by split Cu collars, almost, but not quite, touching the surface to be hardened. High-frequency current is turned into these collars, called "inductors," by an automatic timing device, and very large amounts of energy are thereby induced into the surface to be hardened, heating up the outside shell with great rapidity. The equipment is shown in Fig. 15. It is a characteristic of such heating that the induced current is strongest at the surface nearest the Cu collars. The collars themselves are not appreciably heated.

Operation. The intensity of the induced current is regulated by the design of the equipment, and the time it is allowed to act can be regulated to a split second by suitable electrical devices. Thus just the desired amount of heat can be supplied and the time allowed for the heat to soak toward the center of the piece can be regulated so that only a very shallow depth need be heated above the critical temperature before quenching.

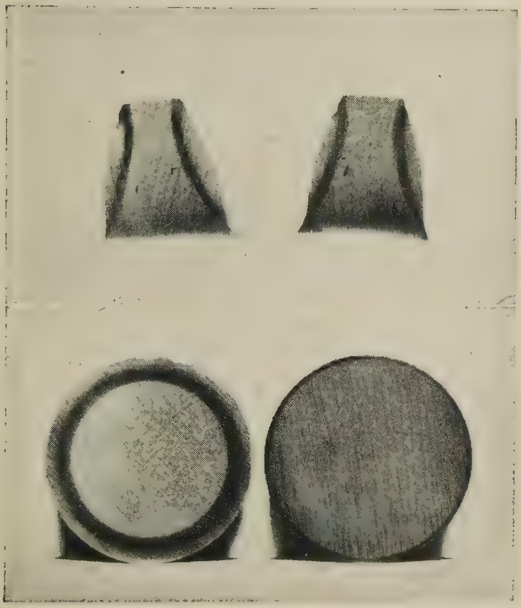


FIG. 14.—Sections of flame-hardened objects. Above, gear teeth; below, 1⅜-in. diameter rounds with heavy and light case. (Courtesy A. B. Kinzel.)

Quenching is accomplished without moving the piece or the collars, since the collars are hollow and are perforated with many small holes so placed that jets of water passing through them will envelop the whole piece without any dead spots. High-pressure water jets are automatically projected through these holes at the proper instant.

Results Obtained. Thus entirely reproducible and thoroughly uniform treatment can be applied to the surface of a symmetrical object.

Figure 16 shows a section of a crankpin bearing. Crankshafts so hardened are coming into considerable use. Boegehold[52] refers very favorably to crankshafts so hardened, from the point of view of a user automotive engineer. He comments that where such treatment is applicable, plain C steel, or slightly alloyed steel, e.g., that with 0.25% Cr, may sometimes be substituted for more expensive alloy steels.

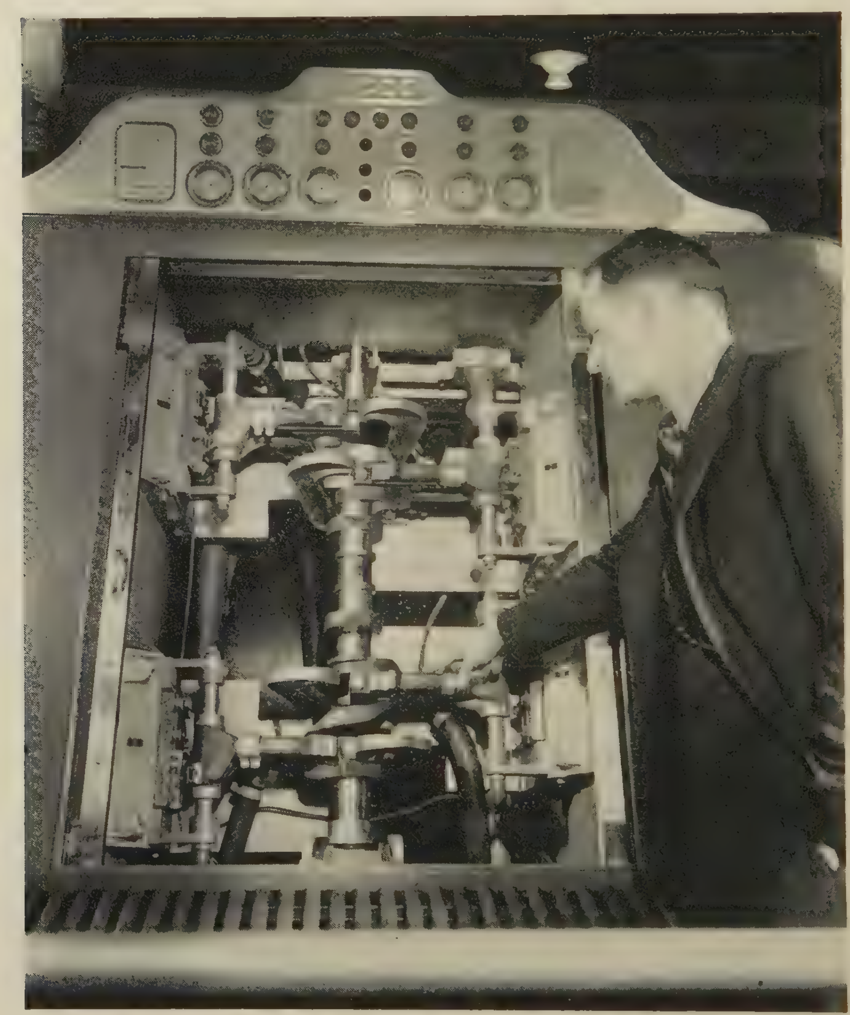


FIG. 15.—Equipment for inductive heating and quenching of crankshaft bearing surfaces. (Courtesy Ajax Electrothermic Corp.)

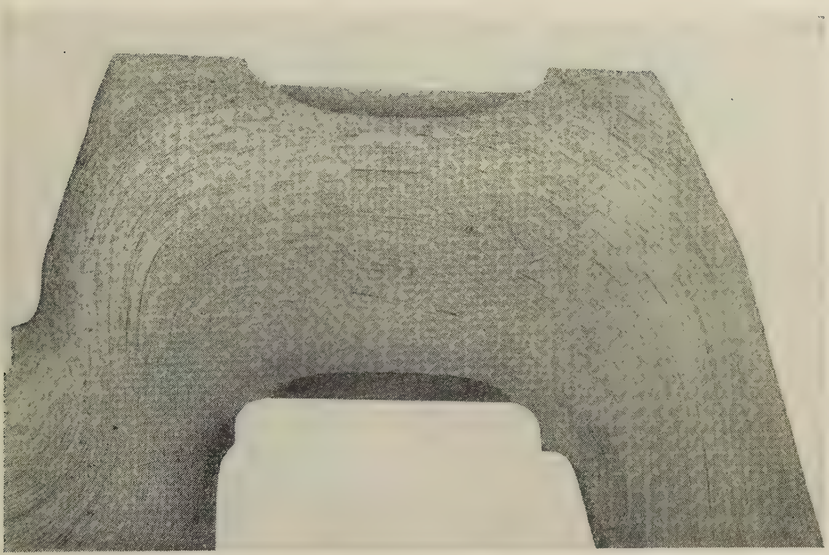


FIG. 16.—Section of inductively hardened crankpin bearing. (Courtesy Ajax Electrothermic Corp.)

General Precautions. Prior to hardening the steel is given the degree of toughness desired in the core and internal strains relieved. Fine-grained steels, normalized, or normalized and drawn, or previously quenched and tempered to a fine sorbitic structure, are usually employed. After hardening, the piece is given a low temperature draw for strain release of the hardened surface.

The steels used generally run from 0.35 to 0.60% C with the lower C on the more highly alloyed steels.

Precise Control Required. The energy used in hardening a crankshaft throw may be of the order of 100–200 k.w. but is applied for only a few seconds, the time being controlled to a small fraction of a second. The heating time depends on the depth of case desired; for most applications heating is accomplished in less than 10 sec.

The hardened surface zone, of say 55–65 Rockwell C, depending on the steel used, may, for example, extend in for 0.10 in., the hardness dropping gradually in the next 0.005 in. to the core hardness of the original material.

The equipment for induction surface hardening is costly; special inductors are required for each job, and the automatic timing has to be worked out for each particular case. The time cycle is so short that the production from the equipment is high; so the method is best fitted

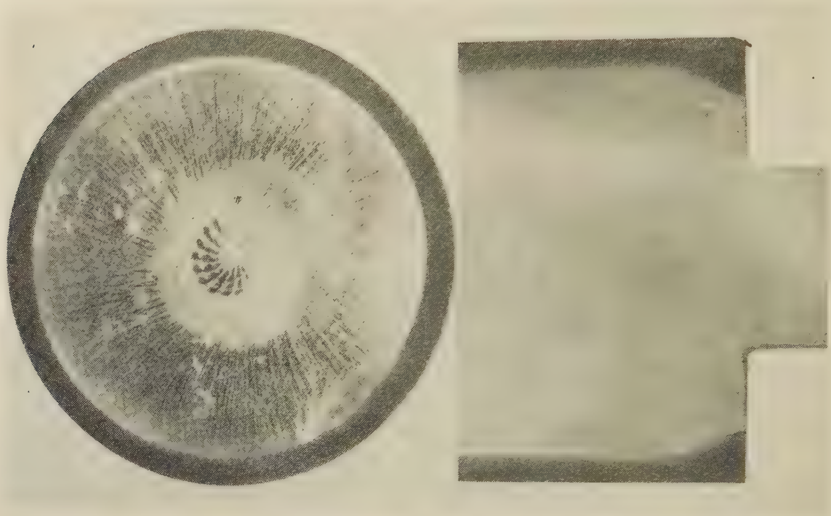


FIG. 17.—Sections of 5-in. diameter roll, inductively hardened. (Courtesy Ajax Electrothermic Corp.)

for quantity production of the same piece. Less elaborate equipment may be used when the shape of the pieces allows.

Roll Hardening. Figure 17 shows horizontal and part of a longitudinal section of a 5-in. diameter leveler roll 30 in. long, made of 0.96 C, 0.52 Mn, 0.19 Si, 0.52% Cr steel.

The roll was pushed through an inductor coil powered with 110 k.w. of 2000 cycle current, at the rate of 1 ft. per 170 sec. The surface was heated to 1450° and held about 20 sec. before reaching the quenching water spray at the end of the inductor coil. The roll was drawn at 350°. The surface hardness was 95 Scleroscope. The gradation of hardness is shown in Fig. 18.

Hard Surfacing. Although the processes are not heat-treating processes, it might be recalled that still other methods may be utilized to provide a special surface on steel, for corrosion-resistant or wear-resistant use. One metal may be coated upon another by electroplating, and an analogous heavy deposit may be made quite rapidly by metal spray.

Coatings of Stellite and of various hard carbides and borides may be applied by welding methods. Worn crankshafts are commercially built up by metal spray [53] and the application of welded-on Stellite facings is general practice for exhaust valves of aircraft engines. Wear-resistant faces or cutting edges of the hard carbides are similarly applied to parts that must resist severe abrasion.

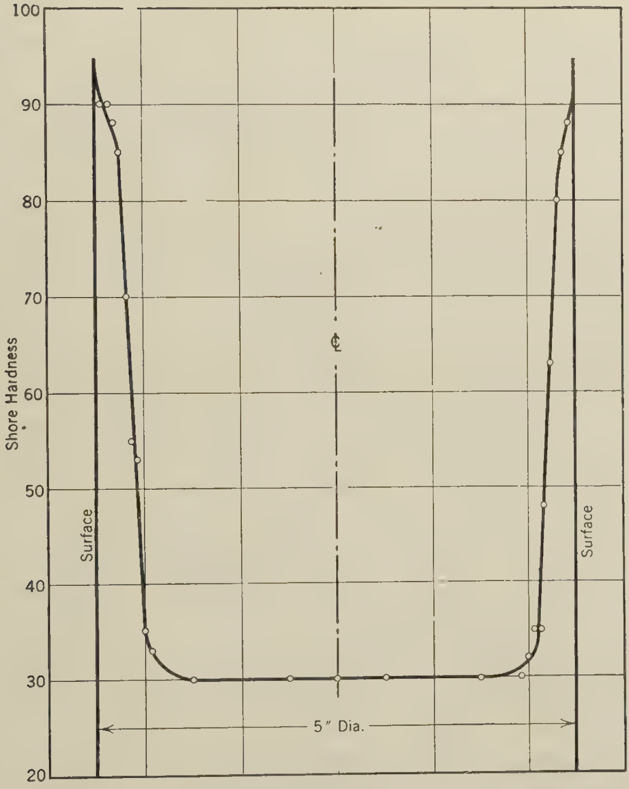


Fig. 18.—Hardness gradient on transverse section of 5-in. diameter inductively hardened roll. (Courtesy Ajax Electrothermic Corp.)

Cladding, Calorizing, and Analogous Processes. Carbon steel plate or sheet carrying a thin skin of stainless steel or of Ni on one or both surfaces, made by casting the steel against a plate of the corrosion-resistant material so that it welds to it and rolling the duplex billet down, or by equivalent methods, has recently become commercially available. This plate is analogous to the composite plates of high- and low-C steel that have long been used for making plowshares and the like. These "clad" steels are fabricated by hot or cold forming and by welding, and suitable means are employed to protect the cut edges. Some irregular objects cannot be made by such methods, but it is, nevertheless, possible to provide objects of almost any shape with special alloy surfaces, by one method or another.

Surfaces high in alloy content can be built up by methods somewhat analogous to carburizing, in which the steel is heated in a granular packing of the metal to be applied and a suitable atmosphere maintained, or by heating in a gas from which the desired metal can be deposited, for instance, a volatile chloride of the metal. After the required thickness of metal has been secured on the surface, a further heating to allow diffusion, is usually required.

Aluminum is added by heating in Al powder plus some inert material to keep the powder from caking. The coating is then given a diffusion treatment and superficial oxidation is allowed, in order to secure a heat-resistant coating. This process is called "calorizing." Though it does not find very wide application, it is useful in specific cases.[54, 55]

Chromium or Si may be added in similar fashion, usually in the presence of a volatile chloride, in the "chromizing" and "siliconizing" processes for production of corrosion-resistant surfaces.[56, 57] These processes have not yet been very completely worked out on the production basis and their economic value cannot yet be appraised. The general principle of using an easily fabricated, cheap steel as the base and giving it a surface of the sort necessary to resist scaling, corrosion, or wear will undoubtedly find increasing application. For the present, at least, these special processes will be carried out in jobbing plants specially equipped and experienced, rather than in the average heat-treating department, so they need not be discussed in detail here.

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SECTION III. CONTROL OF HEAT-TREATING OPERATIONS

CHAPTER 14

PROPER HEATING IN FORGING AND ROLLING PRACTICE

It has been brought out in earlier chapters that heat treatment depends upon control of temperature of the product in heating and cooling. The structure and properties obtained are vitally dependent on temperature. It is of no avail to understand the principles of the time-temperature relations involved in heat treatment unless they can be effectively put into practice.

A heating operation, not ordinarily under the heat treater's control, and one that precedes heat treatment proper, is that of heating for forging or rolling. Both because it exemplifies in intensified fashion some of the problems which the heat treater encounters and because the structure produced in the forging affects its suitability for later heat treatment, we shall discuss this operation.

That the initial structure of a forging may make it an almost impossible task for the heat treater to produce good results has been brought out in Chap. 3. The evil effects of overheating or of attempting to forge with the outside dripping and the inside hard are in strong contrast with the good results from forging a piece that has had time to become thoroughly equalized in temperature, but not overheated, with the temperature so adjusted that forging is completed with the steel not far above its critical range. High heating and finishing temperatures and lack of sufficient working produce structures like that of Fig. 1, and the opposites produce structures like that of Fig. 2. These specimens represent the average work of the same crew, paid on the same basis of output, on the same size forgings. The poor results of Fig. 1 came from a furnace that had to be forced to heat the billets as fast as the hammer men needed them, so that the billets had badly overheated surfaces. When a different furnace was installed, one in which the steel could be given more time to attain temperature uniformity, with a lower furnace temperature, and which could supply sufficient properly heated stock to keep up with the hammer so the crew did not have to force it, the good result of Fig. 2 was produced.

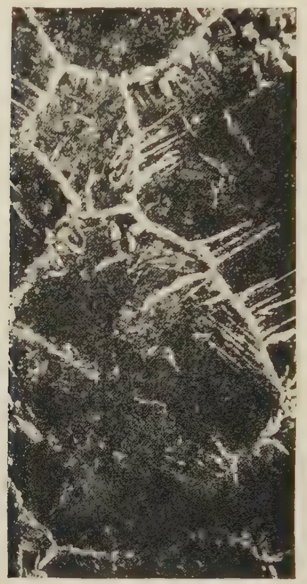

FIG. 1.—Structure of large shaft improperly heated.

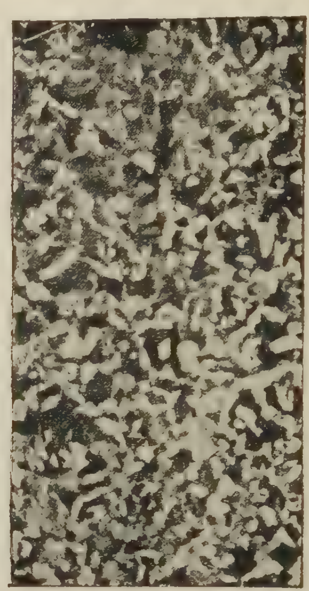

FIG. 2.—Structure of large shaft properly heated.

×100 (Bullens)

Properties Obtained. How such structures are reflected in the response of the steel to heat treatment is shown by the data below, and by the fractures of the corresponding tensile bars in Fig. 3.

Piece No.	Tensile	Yield	Elong., %	R. A. %	Fracture
1	88,000	50,500	14	28.5	angular: dead
2	80,500	50,000	11	13.5	burnt
3	96,750	55,750	22	42	dead: half cup
4	86,000	51,250	27.5	39	dead: angular
5	85,750	51,750	22	34	granular
6	96,000	54,000	12	36.5	granular
7	**81,750**	**54,000**	**32.5**	**65**	**silky: full cup**

All the shafts represented by the seven test pieces were made from the same heat of steel, were of approximately the same size, and were subjected to the same amount of working during forging; and all were treated in the same manner at one time. The structures corresponded to the fractures; 1-6 all showed severe overheating. The structure of 6 is shown in Fig. 4. The structure of 7 (Fig. 5), with a

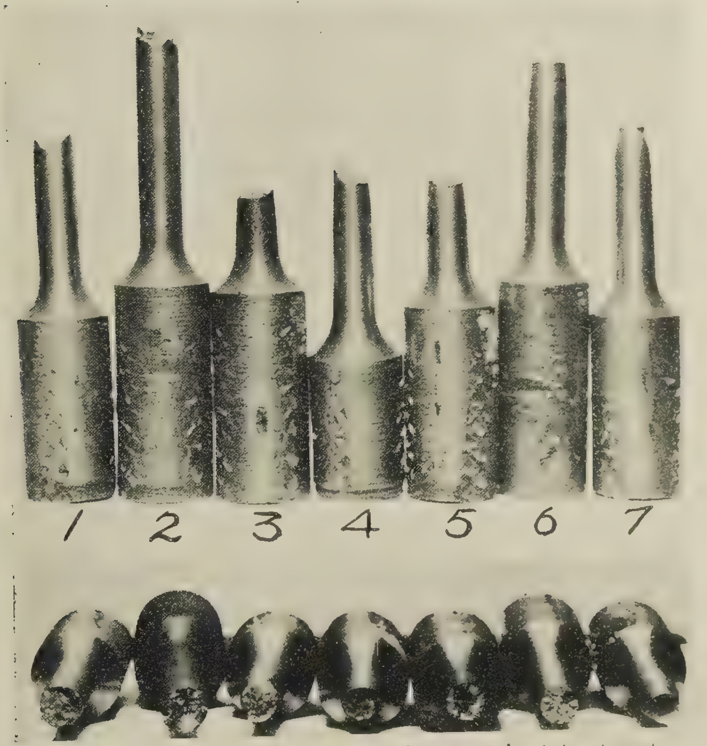

Fig. 3.—Influence of forging on subsequent heat treatment.

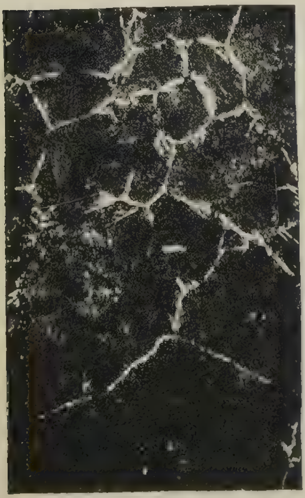

Fig. 4.—Structure of test piece No. 6.

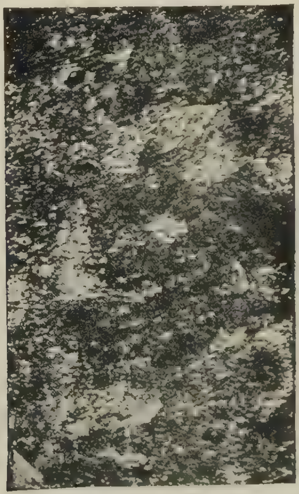

Fig. 5.—Structure of test piece No. 7.

×100 (Bullens)

cup and cone fracture, as well as the tensile results, shows that 7 was properly heated and forged.

Scaling. Slow heating might be expected to lead to increased scaling, but since slow heating allows the use of a lower temperature this tendency is balanced. A gentle circulation of hot gases in the furnace instead of a sharp flame impinging on the work tends to reduce scaling. It is not often attempted to work in a controlled atmosphere since some scaling will occur in the forging operation anyhow, but the effort is to produce a thin, easily detached scale, instead of a heavy one that is hard on dies. Avoidance of too strongly oxidizing an atmosphere will aid in getting the right type of scale. Moreover, as Sherman [1] points out, the heat transfer from a luminous flame throughout the furnaces is better than from a non-luminous one. The luminous flame is a larger source of heat and the furnace chamber is more uniformly heated.

Forging Furnaces. Many shops are finely equipped and well supervised in the heat-treating department proper, but lack of proper furnace equipment and supervision in the forge shop nullifies the efforts of the heat treater. Some comments on the design and operation of forging furnaces are therefore in order.

The selection and operation of suitable heating equipment resolves itself into a problem of determining: (1) the type, size, number, and arrangement of furnaces and the auxiliary equipment; (2) the form of fuel or electricity best adapted to specific manufacturing requirements and distinctive plant conditions; (3) efficient methods of heating, cooling, and handling the stock to be formed; and (4) reasonably comfortable working conditions for the operatives.

It is common to see furnaces hot at one side and cold at another, or hot at the top and cold at the bottom, and to hear complaints on lack of ability properly to heat steel in a furnace in which the burners blast directly against the stock, which may keep the stock nearest the burner cool and heat the pieces farther away. This is a common fault in many oil-fired forge furnaces, particularly those of "homemade" design, although it must be said that the products of many furnace builders are not altogether free from this criticism. Unfortunately, the common method of designing and building a forge furnace appears to be building a box with working opening at one end, placing a hole somewhere in the wall through which a burner is to blast, providing one or more holes in the roof, and calling it a furnace. All this is done without any regard for the generation and proper application of heat to the stock and without provision for the control of the air entering the chamber or of the hot gases leaving it.

The average forge shop is underequipped in the matter of heating area with the result that the furnaces are pushed beyond capacity to the detriment of the steel, the operators, and the manufacturer who pays the bill. The general design of forge furnaces is far below the standard of heat-treating furnaces and is a point too commonly left to the forge man or to a bricklayer. The average forge furnace is designed poorly in the matter of size, structure, control, size of working opening, protection for the operatives, and unnecessary vents—the last-named representing little more than an unnecessary source of waste. Improvement will follow intelligent operation of larger and heavier furnaces of proper design which will pay for themselves and decrease the cost of operation. Such furnaces should be provided with larger heating chambers, arranged to prevent overloading and to permit of uniform application of heat to the stock without local heating.

It ought to be relatively easy to secure uniform temperature from piece to piece in a forging furnace, when the pieces are not piled one upon the other so that one shields the other from the source of heat, a matter which we shall see later is of vast importance in many heating operations.

Precautions for Uniform Heating. With the normal distribution of stock within the furnace it is necessary to provide relatively large heating chambers and to avoid local overheating or direct impingement of flame in oil- or gas-fired furnaces, which are the most common types. Several burners rather than one large one make it easier to avoid localizing the heat. Provision should be made for the circulation of hot gases, but not of the flame itself, about the stock. Gas circulation should be so controlled that the hot gases do not play into the faces of the operators when the door is opened to remove the stock. A useful type of construction is shown in Fig. 6.

Vents should be so placed that the combustion gases are not short circuited from flame to vent without passing around the stock. Nor should the flame itself burn beyond the vents; sufficient combustion space should be afforded inside.

Vents should be supplied with adjustable dampers since greater vent area will be needed while the furnace is warming up than when regularly operating. If it becomes necessary to heat one end only of a very large billet so that the other end must be allowed to project from the door, the space around the billet should be filled up with brick. The spaces between the bricks act as vents and the regular vent should be nearly closed, to keep from pulling in excess air through the joints of the bricks. In general, all the air entering the furnace should be under control, so that there are no openings that cannot be closed tight, not only to avoid thermal losses from the heating of unnecessary air in operation but also in order to delay cooling when the furnace is shut down and thus to increase the life of the refractory lining. Control of furnace atmosphere is another reason for controlling all the air entering the furnace. Most forge furnaces in operation lack this provision, with the result that it is difficult to maintain the soft hazy atmosphere so desirable to minimize scale.

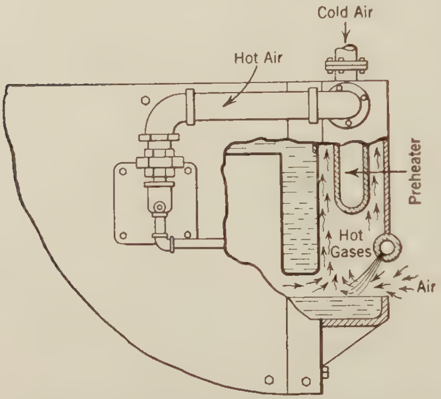

Fig. 6.—Section showing construction for preheating the air and protecting the operator. The hot gases are deflected by downward blast and air is drawn in through the opening into the forge furnace.

Furnace Designs. A wide variety of forging furnace designs are employed, each of which has advantages for specific uses. No one is best for all conditions. Some of these are illustrated diagrammatically in Figs. 7–9.

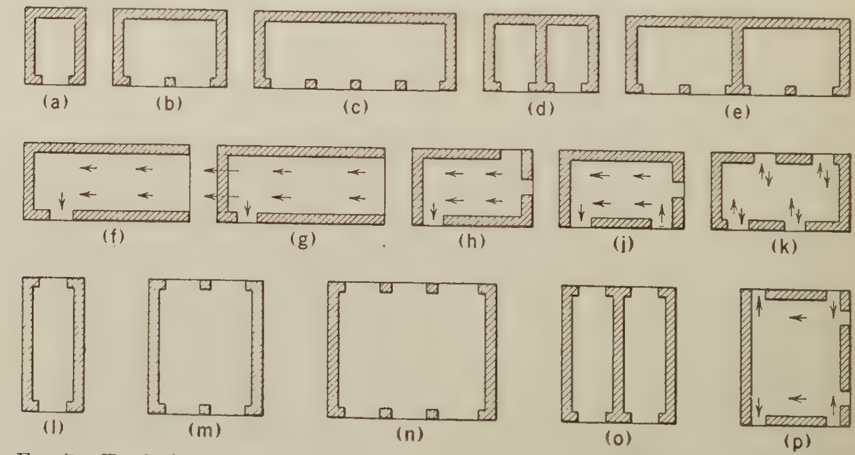

Fig. 7.—Typical arrangements of chambers and working openings in forging and heating furnaces. Types are varied with the nature of the process, with the material to be heated, with the production requirements, the method of handling, and the plant conditions.

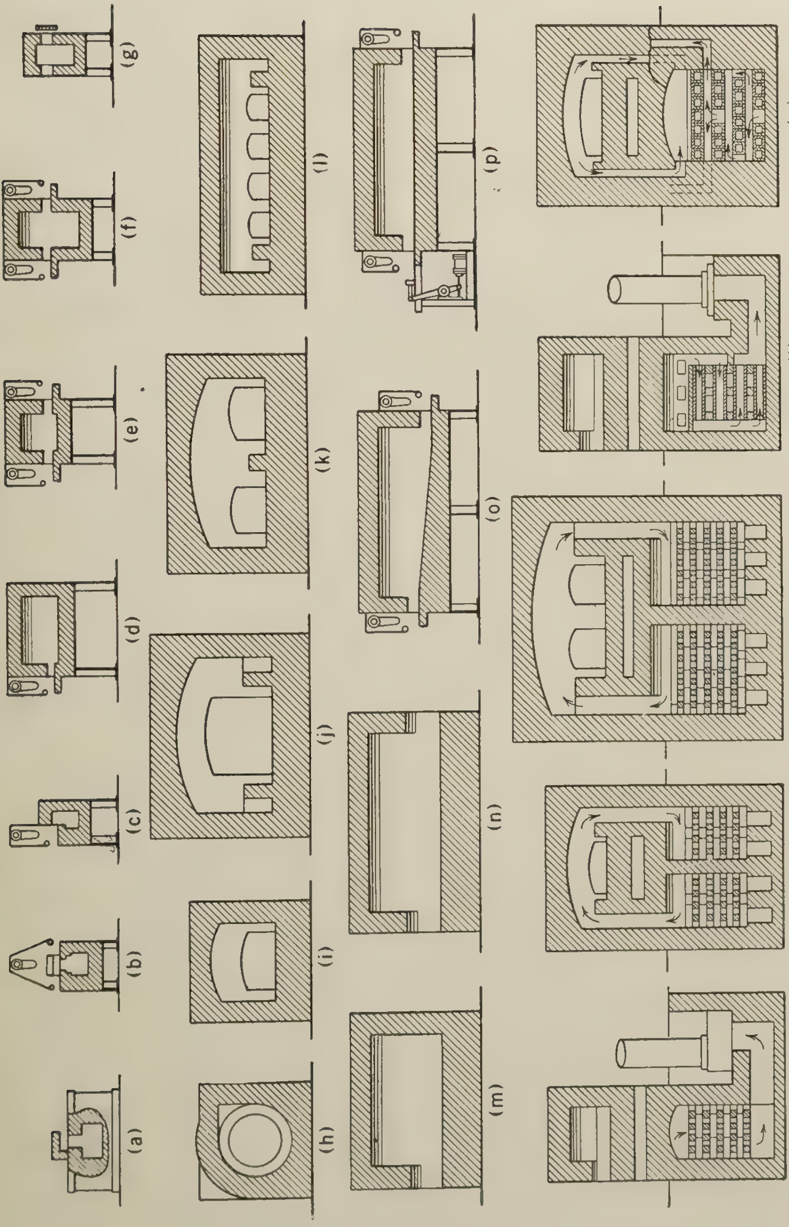

Fig. 8.—Some varieties of heating furnaces.

The mechanical characteristics of the furnace chamber with regard to the method of moving material to, through, and from the chamber; of efficiently utilizing heat within the furnace structure; and of providing comfortable working conditions for the men, are just as essential as the features of the design affecting uniform application of heat to the stock.

The arrangement of the heating chambers and working openings of the furnace proper and the relation of the furnaces to the machine are essential factors affecting the cost of production frequently overlooked in practice. Although there is not as much latitude on this point as there is in the design of the furnace from the standpoint of "heat application," there is, nevertheless, great room for improvement in the general arrangement and layout of such furnace equipment. It is desirable to maintain continuous production, as far as may be practicable, in equipment that may be readily adapted to changing conditions. Frequently, a suitable arrangement of stationary furnaces will permit continuous production without automatic equipment. Figure 7 illustrates, in plan, typical designs of furnaces that have been developed to meet such conditions. The method of generating, applying, or utilizing the heat may vary in any of the designs illustrated.

Figures 8 and 9 illustrate sections of typical furnaces, without regard to the arrangement of the chambers and working openings, and are intended to illustrate some of the different methods of applying and utilizing the heat and of handling material with different forms of fuel.

In Fig. 8, *A* illustrates the ordinary blacksmith forge adapted to coal, oil, or gas, which is probably the simplest form of furnace in use and has its distinct field of usefulness. *B* to *G* illustrate different forms of end-heating furnaces with provision for applying the heat from above or below the stock as required; *H* to *K* different types of furnaces for heating heavy billets; while *L* illustrates a multi-door heating furnace which may be operated by a number of men from one or both sides.

M and *N* illustrate the difference between single-end and double-end furnaces. *O* is a double-end furnace designed to heat pieces that may be readily moved by hand; *P* is a furnace of similar type provided with an automatic pusher.

Q to *S* illustrate different types of regenerative furnaces that have been used for forging, welding, and similar operations, and are designed along the lines of the open-hearth furnace; and *T* and *U* the recuperative type of furnace, designed to utilize the heat in the spent gases to preheat the air for combustion through metallic or refractory recupera-

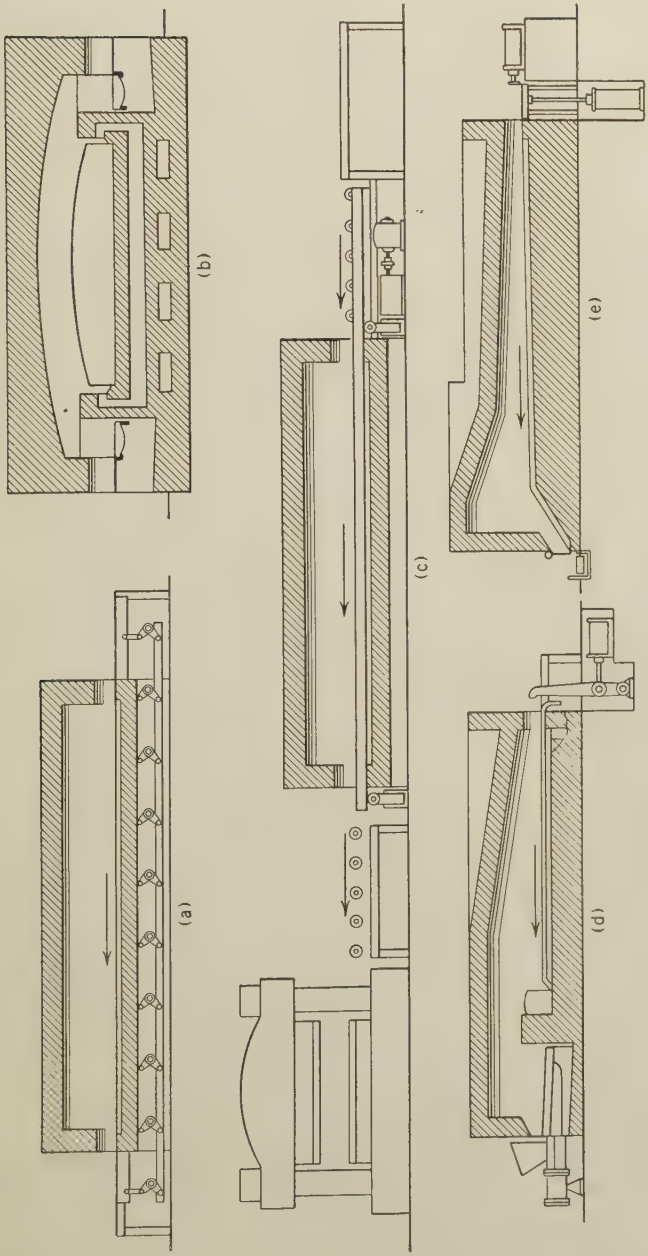

FIG. 9.—More varieties of heating furnaces.

tors. This type is quite popular abroad where the fuel conditions are quite different from those prevailing in this country. As a rule, under intermittent operating conditions, there is a limitation to the efficient use of regenerative or recuperative furnaces, although different conditions influenced by requirements for continuous operation, or the heating of heavy stock in large furnaces, may reverse the conclusions. Regenerative and recuperative furnaces are entitled to consideration, particularly in larger sizes, when it is apparent that there is no simpler and cheaper solution of the problem, all things considered, but enthusiasm for indirect recovery through regeneration or recuperation is frequently unwarranted.

Fig 9*A* illustrates the design of a furnace intended to move the material forward in a series of steps, and *B* an efficient type of plate-heating furnace with relatively wide chamber arranged for coal charged through fire boxes on each side.

Continuous plate heating is effected by the arrangement of press, furnace, and conveyors, illustrated in *C*. The plates are charged, heated, and delivered one at a time. This arrangement is desirable whenever the nature of the product and the production requirements warrant the consideration of continuous methods of production.

D illustrates the stoker-fired type for billet heating, which is built in a great variety of sizes with different arrangements of chambers and working openings. In some cases, particularly with fluid fuels, the discharge end is arranged about as illustrated in *E*.

There is an endless variety of furnaces to meet present-day requirements for better and cheaper heating practice in forging and forming operations. Those referred to above are typical and are intended merely as illustrations of the improvements that are likely to be effected by intelligent study of the individual requirements with due regard to all the factors affecting the quality and overall cost of the heated product and working conditions for the operators.

Slot Heating *vs.* Slugs. Whereas the slot type of furnace is required when only the end of the piece is to be upset, and may be necessary in the case of a restriking operation that involves reheating, one may question the suitability of slot furnaces when the nature of the work allows the use of slugs.

The practice in which the bar stock is inserted in the slot or in holes, the end heated up, sheared off in the first dies, and the remainder of the bar stuck back into the furnace makes for hot working conditions and for low fuel efficiency. Any furnace designer well knows that "through metal" reaching from furnace chamber to the outside is to be avoided and that losses through open doors are so huge that quickclosing doors are necessary. The slot furnace operates with through metal or with open slots all the time. Slot openings should be adjusted in height and width to suit different sizes of stock; also the opening should be as completely filled with stock as possible at all times.

This adjustment may readily be effected by laying a course of loose brick on the sill; and, when necessary, laying a loose brick at the side of the opening.

Unnecessarily high openings are responsible for high fuel consumption, unnecessary heat on the operator, fumes in the shop, and oxidation, particularly when the burner blasts through an uncovered hole. The fuel consumption is more affected by the size of the opening than by the burner.

Proper practice in slot furnaces is to take out the hottest piece and replace it by a cold one—i.e., "one in and one out," rather than to do "batch" heating. That is, the sequence should be such that each piece is in the furnace the same length of time.

Slot practice is defended on the score of supplying a tong hold; yet the cost of heating will generally overbalance the cost of the extra metal necessary for a tong hold when the charge is cut into slugs and passed through a closed furnace. One up-to-date plant uses slugs, the first forging operation being in a press that elongates and reduces one end to form a thin tong-hold without wasting stock. The result is a real saving in manufacturing cost.

One plant is operating on 3 in. × ¾ in. stock formerly heated in a slot furnace, a heated length of 5 in. being required. With slot practice they used 50 gal. oil per ton. By shearing to 5-in. lengths, and heating in the same furnace, merely providing an uninsulated boiler plate door over the slot and keeping the slot closed except when charging and discharging, the oil consumption was cut to 35 gal. per ton. With a pusher type or other continuous furnace, properly insulated, the fuel consumption could, of course, have been cut still more.

Fuel Consumption. Fuel consumption alone is no criterion of the usefulness of a forging furnace. Were it necessary to use more fuel in order to secure properly soaked and uniformly heated stock, it would be economy to do so. However, properly designed furnaces with ample heating space operating at a moderate rather than an excessive temperature and with proper flow of hot gases seldom use as much fuel as the poorly planned furnaces they replace.

Delays. A special difficulty attends the operation of furnaces that operate at an excessive temperature when a hammer hold-up occurs and the stock has to remain in the overheated furnace. Even shutting the fuel off entirely may not prevent "burning" of the piece and coarsening of its structure. Automatic control of the ruling temperature of the furnace, through pyrometers, is helpful in continuous furnaces of the pusher type for example, and indicating pyrometers are a great help in manual control; but it is rare indeed that nothing occurs to the flow of work through a forging furnace in the course of a day temporarily to vitiate the effectiveness of automatic control on the temperature of the stock itself.

Mitigation of Scaling. Scaling in heating for forging may be mitigated by adjusting the fuel-air ratio to give "soft" heat; but, when full atmosphere control is desired for entire prevention of scaling, electric heating may be resorted to, and a suitable protective gas led around the work. However, this will prevent only the scale formed during heating, not that which is formed during the actual forging operation.

Electric Heating. In the case of stock of uniform, and not too large, cross section, heating can be done as in the Berwick type heater [2] by passing an electric current through the work itself, which not only allows rapid heating without the cold core that accompanies rapid heating from an external source but also, since the pieces are individually heated, allows an "electric eye" automatically to shut off the current when the desired temperature has been reached and, then, if the piece is not withdrawn at once, and so drops below the proper forging temperature, to turn it on again. Hammer delays thus exert a minimum of detrimental influence on the heating process. Unfortunately, the range of stock that can be so heated is limited; hence truly automatic control is not widely applicable to all heating for forging. Figures 10-12 illustrate the Berwick heater.

Induction heating by high frequency current offers similar opportunity for individual heating and automatic control.[3] The cost of high frequency generators limits the application of this method. Because of the high temperature, neither radiant tubes nor the ordinary NiCr resistors will survive in forging furnaces. Globar (silicon carbide) resistors may be used. The problems of control of the actual temperature of the work and the distribution of the source of heat so as to get temperature uniformity in the Globar-heated furnaces are analogous to those in a gas- or oil-fired furnace. Since the resistor life is longer the lower the temperature of operation, there is an added incentive in such furnaces to hold the maximum temperature down, which is to the benefit of the work. This type of furnace is in use, but is rather rare, as are installations of induction heating.

The Berwick type heater is less rare, indeed, it is very common

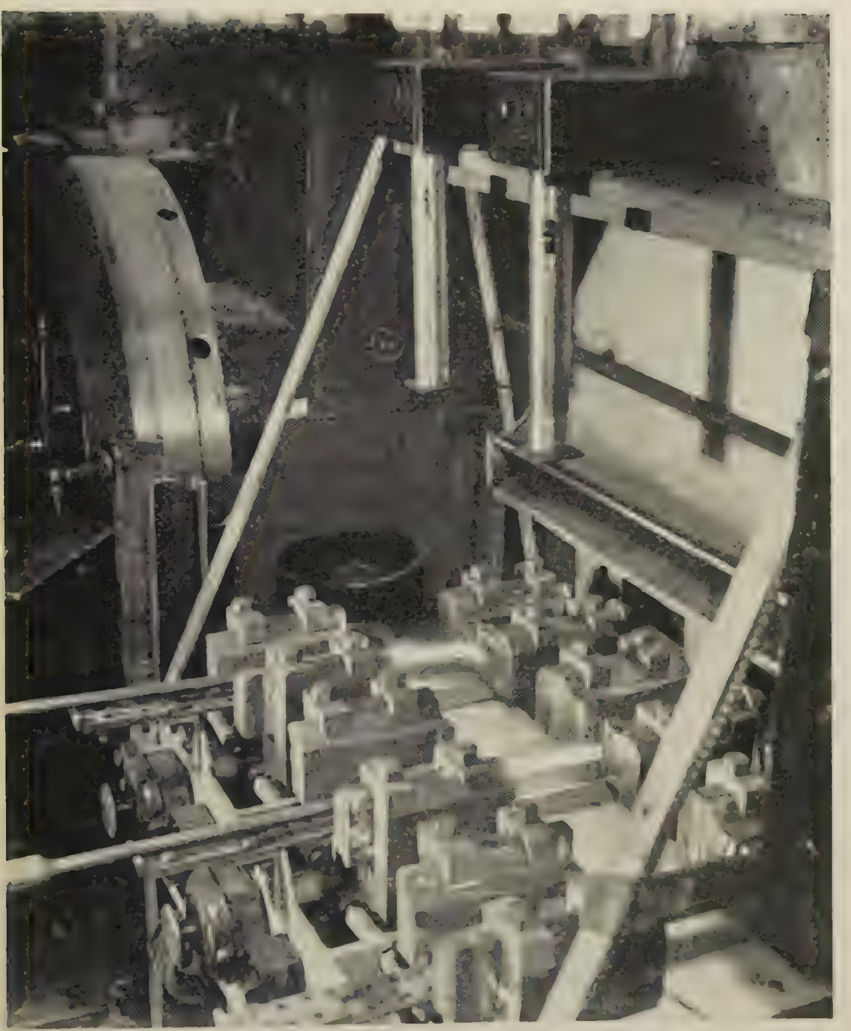

FIG. 10.—Manual feed Berwick electric forging heater in operation at Chevrolet Motors Corp., Bay City, Mich. (Courtesy American Car & Foundry Co.)

as a rivet heater, but it does not cover a large enough section of the field to vitiate the statement that, in general, no heating operation on solid steel depends as much on the "human element," the judgment of the heater, as does heating for forging.

Heating for Rolling. Practically all the comment made above on forging is equally applicable to heating for rolling, though the conditions of uniform flow of stock from the furnace to the mill are generally

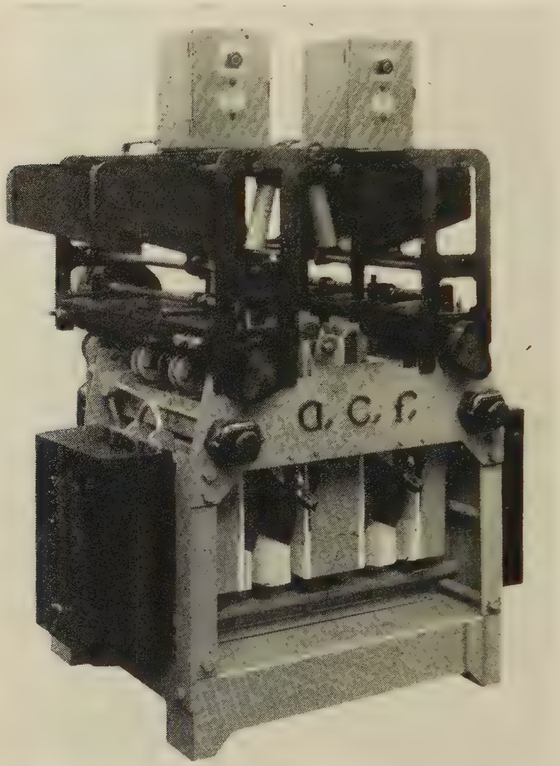

Fig. 11.—Automatic electric forging heater. This is a double unit.

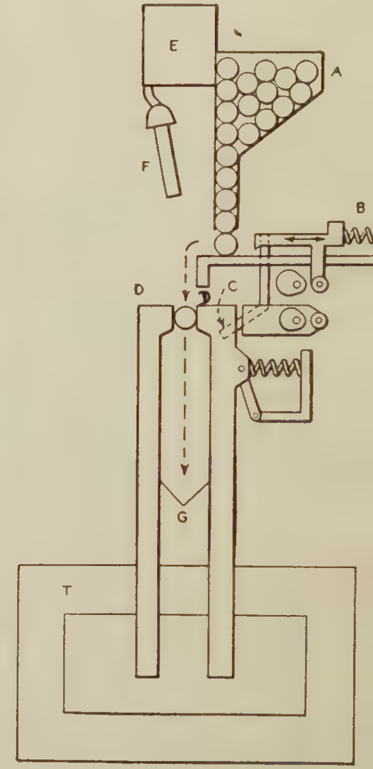

Fig. 12.—Diagram of automatic feed.

Note the "electric eyes" for temperature control. (Courtesy American Car & Foundry Co.)

more favorable than to the hammer or press. However, an improvement in the uniformity of as-rolled stock that would often eliminate the necessity for normalizing prior to heat treatment might frequently be made in rolled products by better control of heating.

Personnel and Supervision. Proper heat treatment of steel is not possible without proper heating in the rolling mill and forge shop, and forgings cannot be produced as they should be without good men and good furnaces. Heating is a skilled occupation. "Any kind of a man" is not good enough for the expensive operation of forging and heat treating.

This factor is the weakest link in the chain of heat-treatment operations and must be strengthened, not only because it is necessary from the technical side but also because it is profitable from the commercial side. The time and money which can be saved in scrap, fuel, power, repairs to furnaces and other equipment, machining operations, lost business owing to defective material, etc., should certainly be sufficient inducement to bring about the change, the necessity for which has been too long overlooked and the possibilities too little realized.

There should be continual supervision of the heating operation in order to check the temperature, the time of heating, the uniformity of heat throughout the chamber, the amount of waste gases, the oxidizing effect of the furnace atmosphere, the consumption of fuel and power, and the repairs to furnaces and other equipment. Personal supervision is made absolutely necessary by the very nature of the work. Pyrometers and the like are desirable and necessary, but vital errors may creep in, even though such appliances indicate they should not exist.

The need is for a better understanding of the cause of present-day conditions and of the factors that control the effort for improvement; and for operators or supervisors qualified to apply properly in practice the simple principles governing the conduct of such heating operations.

The practice of delegating one or more men to supervise the work of heating by checking the points above enumerated, under the direct charge of the shop foreman, is steadily gaining.

Summary. Concisely, the trend of practice should be toward lower furnace temperatures—as close to the temperature actually required in the metal as is consistent; reasonable loading with due regard to the method of exposing the individual piece to the heat so that it will be subjected to the same temperature for the same time and in the same manner as every other piece (as will be discussed for heat treating in general in Chap. 16); lower rate of heating, to permit of thorough saturation; furnaces that permit of better application of heat to the metal and better utilization of heat within the furnace; control of all the air entering the furnace; better working conditions for the operatives; and a watchful regard for the effect of the human element, indicated in terms of capable, trained operators and competent supervision. All these factors contribute to better quality and greater production at less total cost in terms of finished product.

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CHAPTER 15

SOURCES OF EFFECTIVE HEAT FOR HEAT-TREATING OPERATIONS—FUELS AND FURNACES

The suitability of a fuel for heat treating depends primarily upon its ability to be controlled, so that the supply of heat may be continually adjusted to the needs of the charge. The burning of solid fuel on a grate or a stoker is difficult to adjust to these needs. Save for relatively few cases where precision of results is not aimed at, fluid fuels are preferred to solid fuels for heat treatment. Pulverized coal burners could be adapted to some operations of heat treatment, but the deposition of fly ash on the work would have to be avoided in a direct fired furnace. Most modern heat-treating furnaces are gas or oil fired, or electrically heated. This fact itself shows that mere cost of a B.t.u. absorbed in the work is not the main criterion for choice, since coal firing would generally be cheaper on that basis alone. That a snap judgment as to selection of fuel may not be a correct one will be indicated by the listing in Fig. 1 of some of the factors that may be involved.

Fluid Fuels. Confining ourselves hereafter to the fluid fuels, oil, gas, and electricity (since the last flows along wires), we find all these to show flexibility of application and control.

Composition of the gas is an important factor in the matter of selection. For instance—producer gas, water gas, city gas, and natural gas, have about the same control characteristics. However, the difference in composition frequently affects the choice. In certain forging operations city gas is not as desirable as some hydrocarbon fuels because of a difference in composition. The difference is frequently offset by maintaining the so-called "luminous flame." A difference in the percentage of sulphur might make one gas preferable to another.

It is the composition of the combustible mixture and the products of combustion, rather than the composition of the fuel itself, that are controlling factors.

Propane and butane are used much more extensively for heat-treating operations than is generally realized, and with excellent results. Among the fuel gases, natural gas is the usual choice in localities where it is available. Manufactured or "city" gas and mixtures of this with natural gas, water gas, and producer gas are all usable. Because of the differences in B.t.u./cu. ft., burner design will vary according to the gas used. Recuperation to preheat the air for combustion may be necessary with producer gas or other low B.t.u. gas.

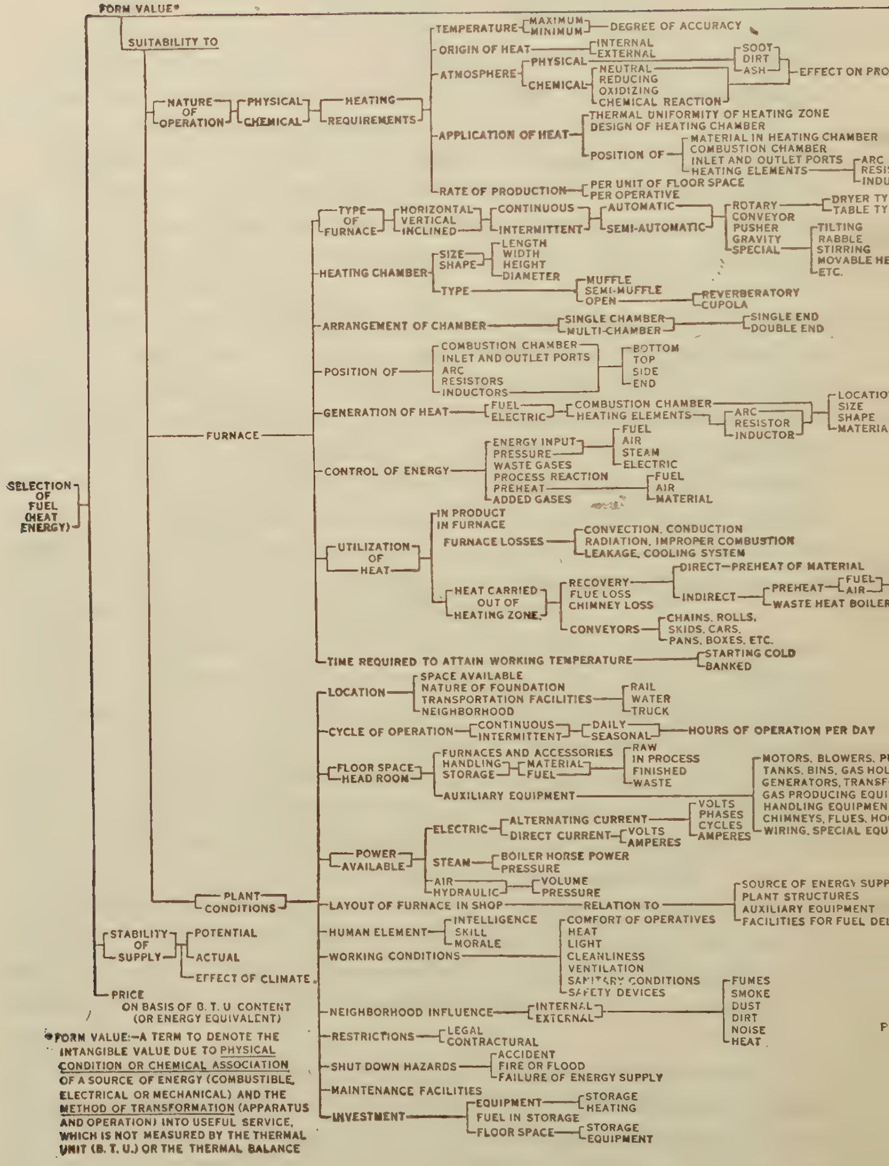
Propane and butane may be shipped as liquids in cylinders or tank cars under pressure and stored under pressure. By release of pressure the liquids gasify and the gases are then handled like any other gas. Gas firing may thus be used where natural gas is not available and city gas too costly.

Efficiency and Effectiveness. No cost comparison can be made on the B.t.u. basis alone. The efficiency of the fuel in the particular furnace to be used and the actual manner of placing the work in it have to be considered also. Figure 2 takes in both factors. Even such a comparison is often illusory, for a furnace and charging method that are "efficient" in the use of fuel but do not produce the necessary uniformity of temperature in the work itself, thus leading to rejections, might be very much more expensive to operate per unit of acceptable output than a less "efficient" furnace, on the thermal basis, but one that actually does the job it is necessary to do.

"Effective" might be a better term than "efficient" to use in discussing fuels and furnaces for heat-treating use, since the word efficiency is ordinarily used merely to express the ratio between thermal units supplied and thermal units absorbed by the work. Recalling the previous chapter on heating for forging it is seen that, though the total thermal units in a piece with an overheated surface and an underheated interior might be just the same as in one properly heated, the quality is quite different in spite of the quantities being indentical.

Although thermal efficiency of a heat-treating furnace is entirely secondary to its ability to heat the stock uniformly with as little scaling as possible, there is no reason to avoid those means by which thermal efficiency and greater output per man and per furnace might be obtained.

Internal Heating. One principle of the efficient application of heat which is seldom given much consideration in heat-treating furnaces is the principle of internal heating, i.e., locating the source of heat inside the charge so that radiation is direct to the charge rather than to the walls. This is employed to some extent in the inside resistors for bell-type annealing furnaces for coils of wire or strip (Figs. 3 and 4) and in a somewhat analogous location of radiant tubes in some large annealing boxes. (Figs. 5, 6, and 7.)



*FORM VALUE:—A TERM TO DENOTE THE INTANGIBLE VALUE DUE TO PHYSICAL CONDITION OR CHEMICAL ASSOCIATION OF A SOURCE OF ENERGY (COMBUSTIBLE, ELECTRICAL OR MECHANICAL) AND THE METHOD OF TRANSFORMATION (APPARATUS AND OPERATION) INTO USEFUL SERVICE, WHICH IS NOT MEASURED BY THE THERMAL UNIT (B. T. U.) OR THE THERMAL BALANCE

Fig. 1.—Factors governing selection o

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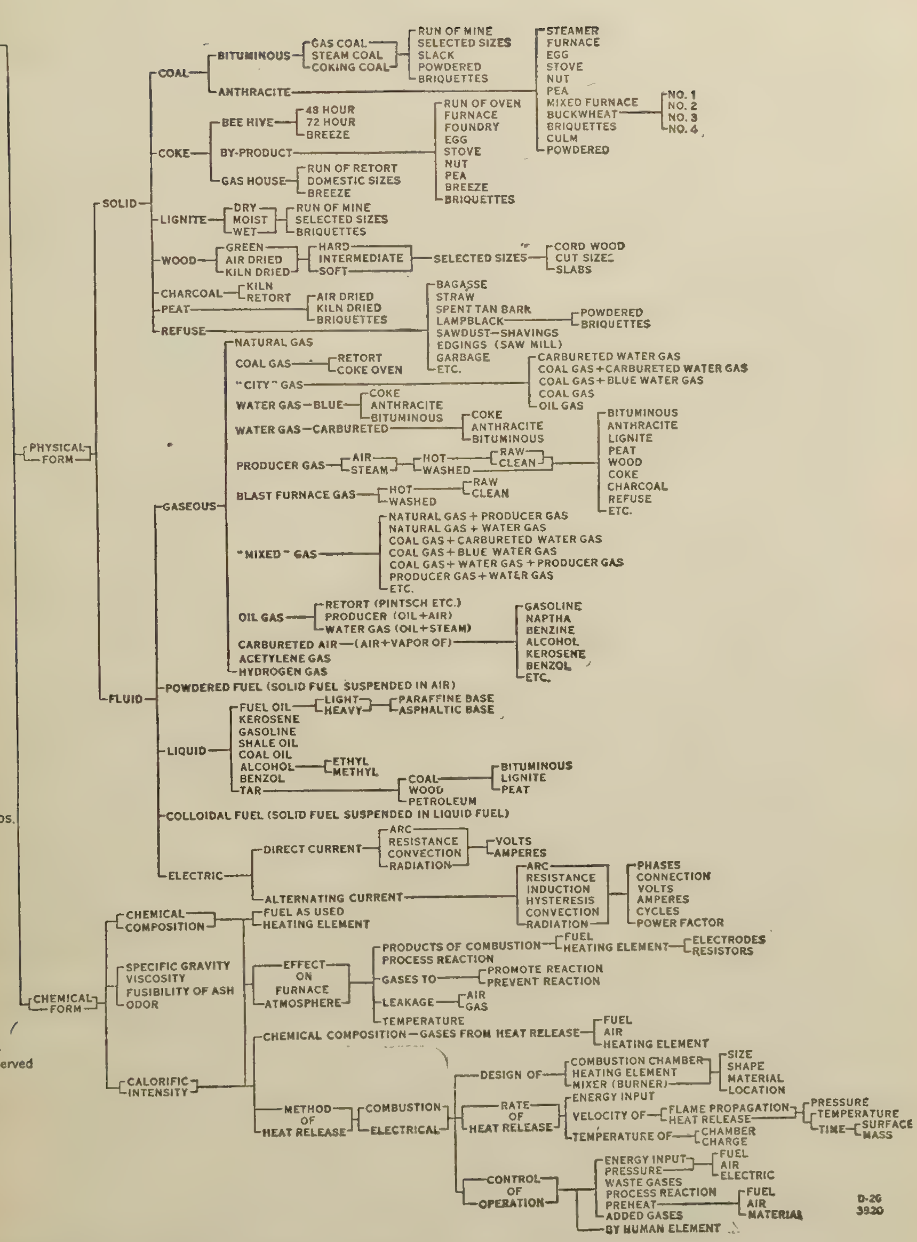


at energy) for industrial heating operations.

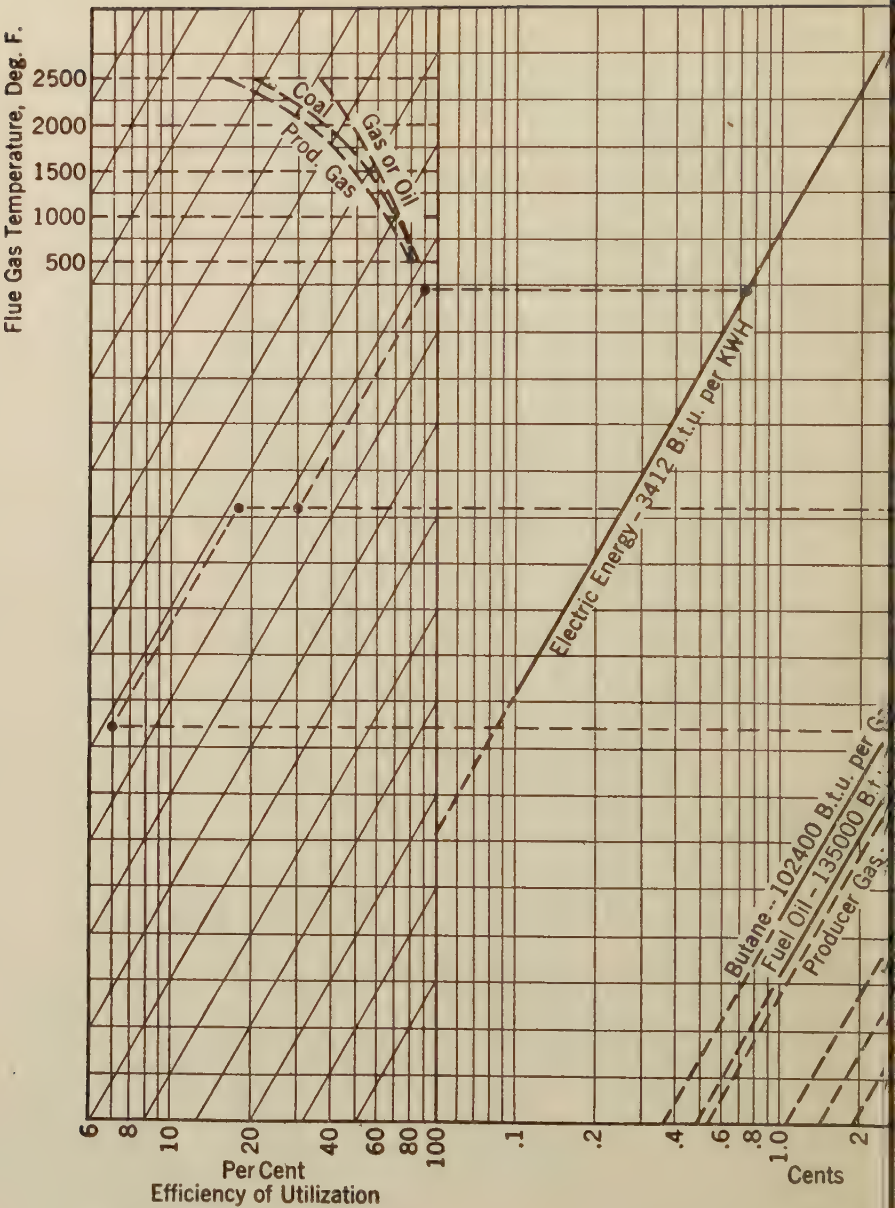


FIG. 2.—UNIT PRICES

Gases, Cents per 1000 Cu. Ft.

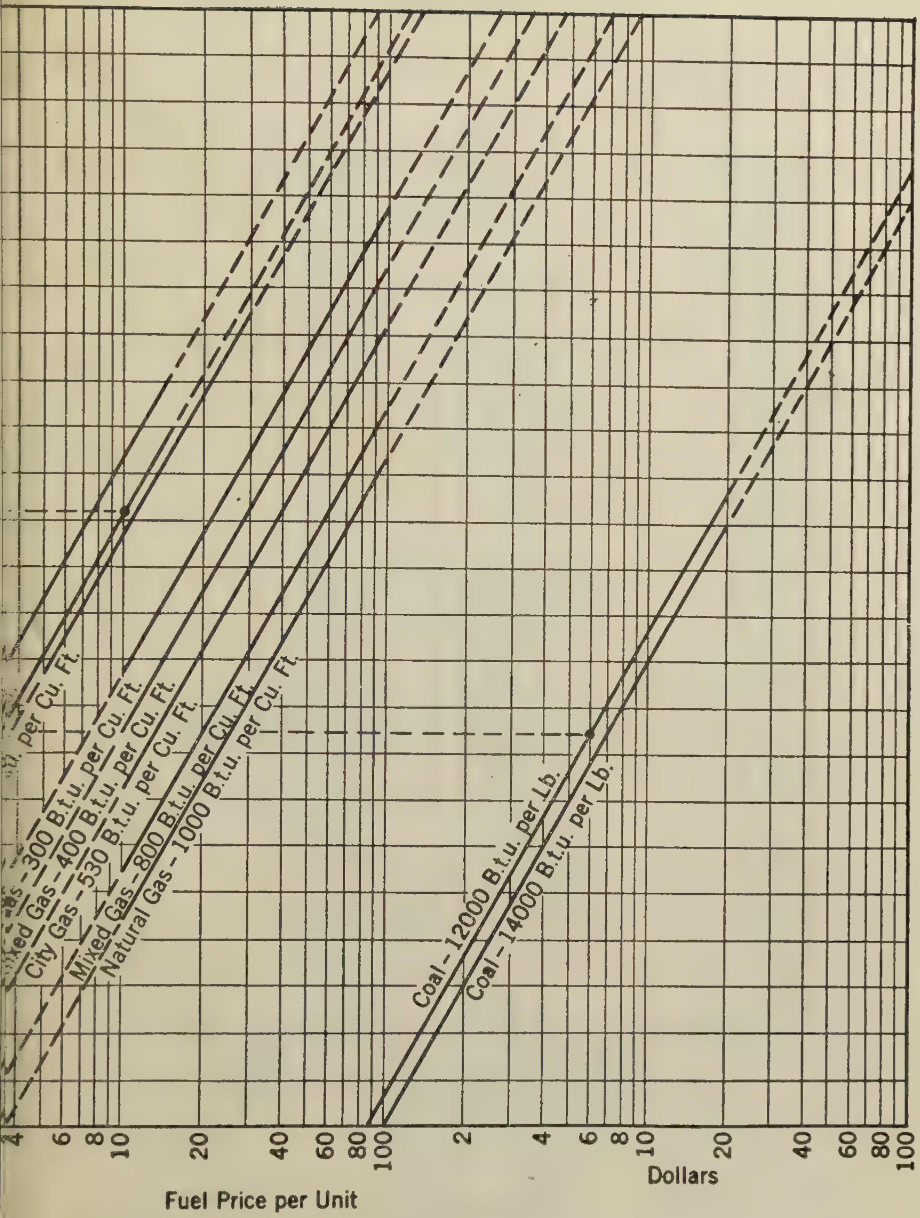
Oil and Butane, Cents per Gal.

Electric Energy, Cents per KWH.

Coal, Dollars per Ton (2000 lb.)

To find the equivalent cost of electric
intersection of the price line and the fuel
of the electric energy (90%). From there
equivalent price (.75c) on the top or bott

The curves in upper left corner show
fuels at various outlet flue gas temperatu
and heat in work and involves other losses



ILLUSTRATIVE USE OF CHART

gy at 90% efficiency compared with fuel oil at 10c per gal. and 30% efficiency. From the ne follow left to the oil efficiency (30%) then follow the diagonal guide lines to the efficiency ow the horizontal guide lines to the intersection of the line for electric energy reading the cale.

oximate percentage (using the common percentage lines) of available heat from the several "Efficiency" in the main chart refers to the total overall proportion between fuel input for instance wall radiation in addition to flue gas losses.

(Courtesy W. S. Rockwell Co.)

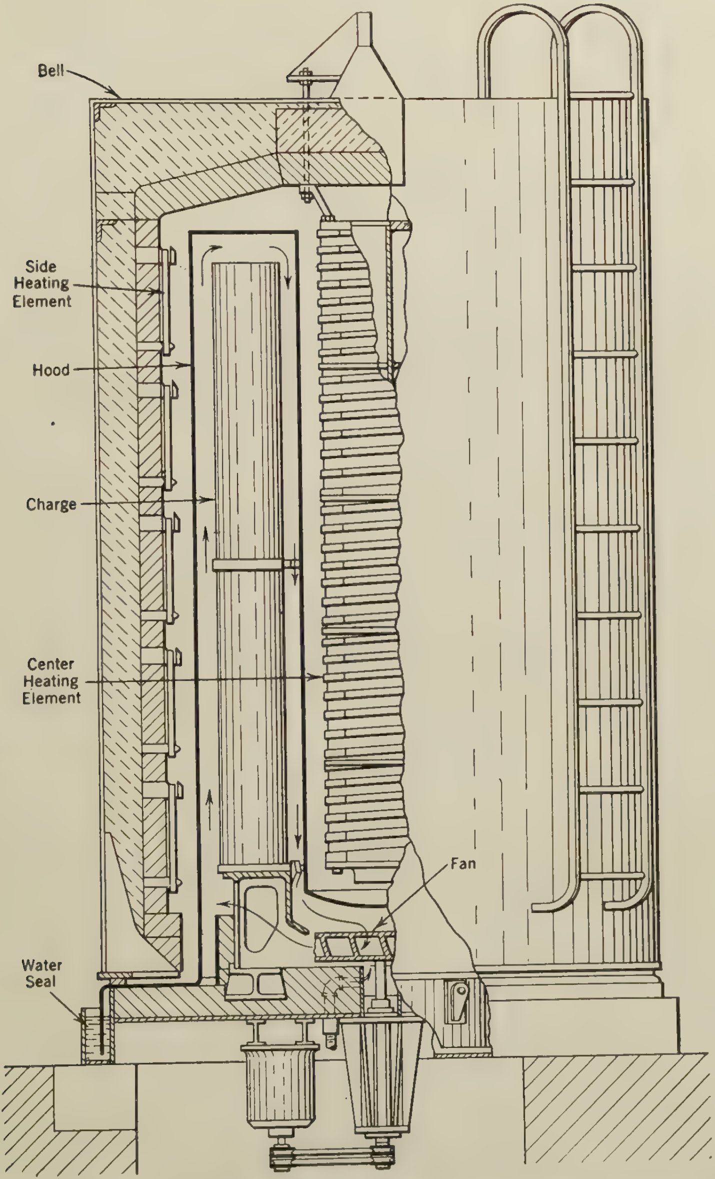


FIG. 3.—Controlled atmosphere, bell-type annealing furnace for steel in coils. Note the fan in the base for circulation and the location of the electric resistors, not only at the sides, but also in the middle of the charge, to give uniform heating. (Courtesy Westinghouse El. & Mfg. Co.)

Bringing the heat into the charge itself by forced circulation of the atmosphere, in some modern annealing and tempering furnaces, and some electrically heated gas-carburizing furnaces of which that of Figs. 8 and 9 is an example, is a big step in the right direction. Various other makes, beside those illustrated, use similar principles. Indeed, one may state that no tempering furnace is modern without forced circulation. Analogous to this is the direct generation of heat within an electrically heated salt bath, which circulates about the pieces being heated, making not only for uniformity of heating but also for thermal efficiency. A diagram of such a furnace is shown in Fig. 10.

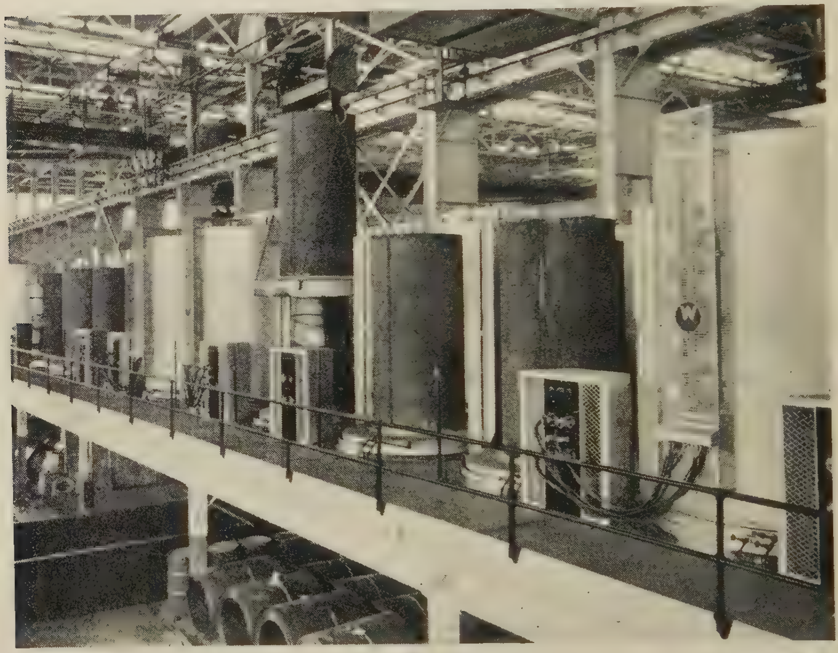


FIG. 4.—Battery of bell-type furnaces of the design shown in Fig. 3, installed at the plant of the Ford Motor Co. Each bell takes two coils of 52 in. wide sheet. Each coil weighs 16,000 lb. The furnace battery consists of 37 bells and 111 bases. (Courtesy Westinghouse El. & Mfg. Co.)

Insulation. The use of insulation on the outside of furnace walls to retain the heat inside for efficiency, as well as to make it more comfortable for the operator, has long been appreciated and is widely used on furnaces for continuous operation. However, there is less advantage in insulation of batch furnaces for intermittent operation since the insulation raises the temperature of the outer faces of the lining and thus causes it to store more heat. By the use of insulation it is often possible to reduce the thickness of the refractory (but poorly insulating) inner lining. Such a change in construction reduces the heat-storage capacity and allows rapid heating of the furnace to its operating temperature.

A recent improvement in furnace refractories has been the development of porous fire brick, refractory enough to stand the furnace chamber temperature, and used without a dense refractory lining. As Bradley[1] and Trinks[2] have pointed out, such brick weigh roughly but one-third as much as an ordinary brick and hence store but one-third the heat. A furnace lined with them comes to operating temperature very quickly.

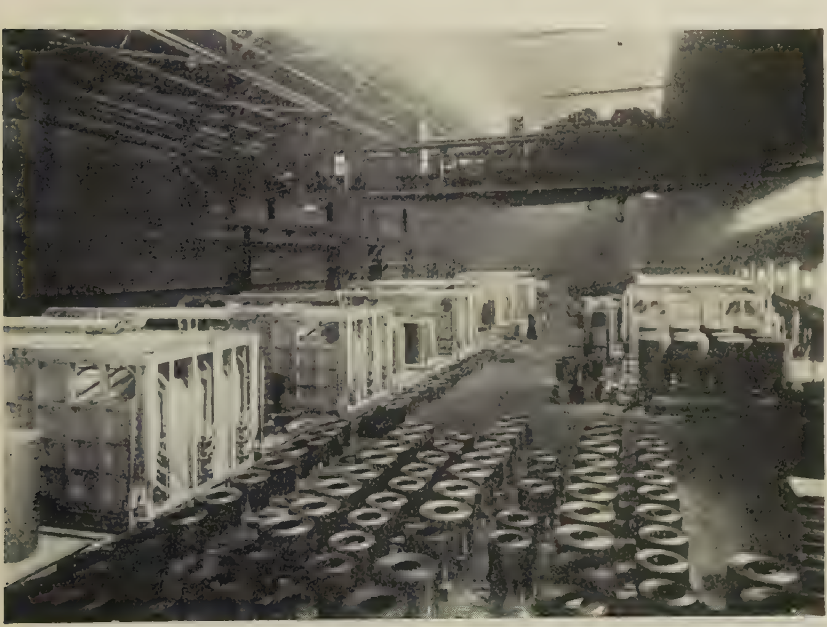


FIG. 5.—Radiant tube annealing covers for coiled sheet. Note the three rows of heating tubes. (Courtesy Surface Combustion Co.)

As an example, Keller[3] cites, for 1600° working temperature, the reaching of that temperature in the furnace chamber, with the same fuel supply, in 24 min. in a furnace with 4½-in. porous refractory brick walls, when with 4½-in. walls of dense brick it took 1 hr., 43 min. Similarly, in furnaces with 9-in. walls, the porous refractory-lined furnace came up in 35 min., that with solid brick walls in 2 hr., 37 min. Whenever the stored heat in a furnace exceeds the loss through the walls, as is generally the case with intermittently operated furnaces, the porous refractory makes a real saving in fuel and in operating time; and, of course, this holds good whatever fuel is used.

These refractories are somewhat more subject to spalling than is dense firebrick, but several makes are on the market that are satisfactorily spall resistant.

Heat Transfer. The principles of heat transfer and of furnace design and construction are worth studying by the heat treater, not merely from the point of view of furnace efficiency but also from that of the light they throw on the effect of different methods of loading, which will be discussed in Chap. 16. Both the furnace problem as such and the effect of different methods of loading are discussed in detail by Trinks[4] and by Mawhinney,[5] and the engineering mathematics of heat transfer is set forth by Schack.[6] The topic may profitably be followed further in such text-books.

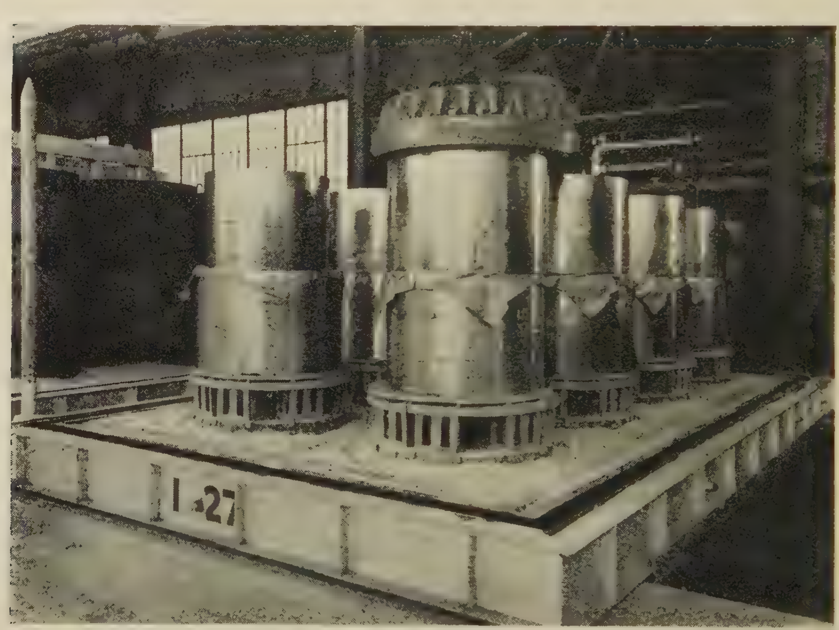


FIG. 6.—Coils arranged on furnace base. Note clearance between rows to give space for the middle group of radiant tubes. Note the open bases to permit gas flow. Each base is provided with a fan, to circulate the controlled atmosphere which is used to prevent oxidation. (Courtesy Surface Combustion Co.)

Distribution of Source of Heat. By use of small, properly spaced burners or resistors, properly distributed, the source of heat may be subdivided and the heat flow may come from a considerable area rather than from a localized one. A diffuse heat source tends toward

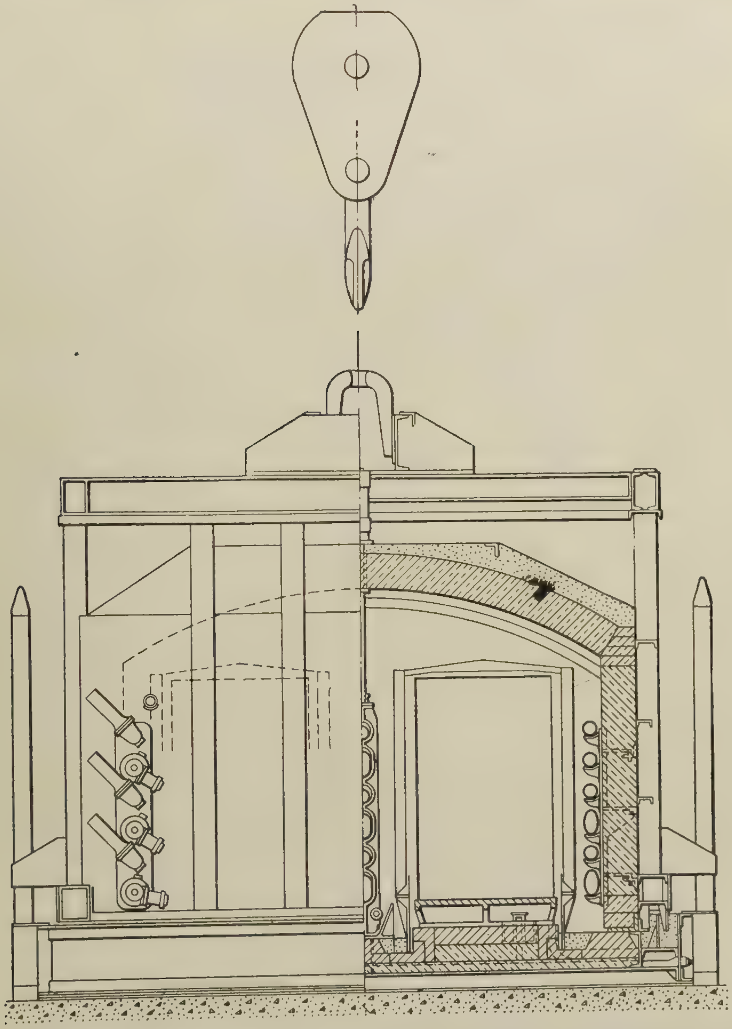


FIG. 7.—Diagram of annealing covers of furnaces shown in Figs. 5 and 6. Note the central row of radiant tube heating elements for better distribution of heat. (Courtesy Surface Combustion Co.)

greater uniformity of temperature throughout the furnace chamber. At high temperature, heating is chiefly by radiation.

Radiation. Obviously it is easier to get heat to a charge by radiation from a number of small sources than from a single larger source.

The ability of the type of furnace used to subdivide its heat source is quite a factor in its suitability. The optimum condition for temperature uniformity is when the whole furnace inside is itself at a uniform temperature, and the whole is a direct source of radiation. This is, save for openings for charging and discharging, accomplished in a muffle furnace, and helpfully approached even in the semi-muffle type where the muffle walls, perforated if the temperature is such that conduction and convection play an important part, are interposed between the hot gases and the stock, to serve as a more diffuse and lower temperature heat source. However, the fuel consumption in forcing heat through a muffle wall and the maintenance of muffles are drawbacks. (It will be shown in the next chapter that getting a uniformly, radiating source is only part of the battle, since the charge must also be placed so as to receive the radiation uniformly.)

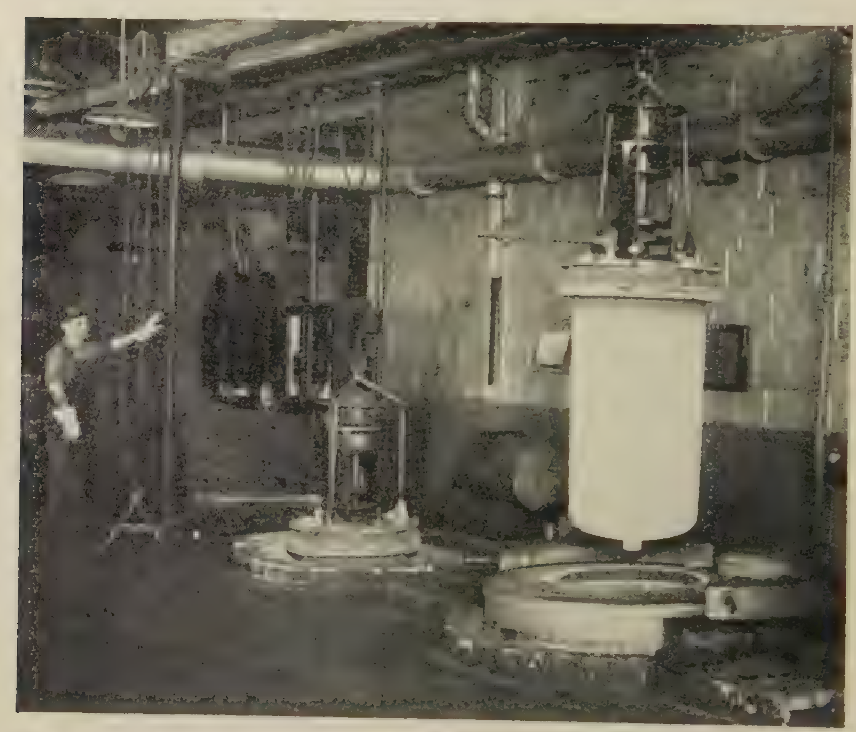


Fig. 8.—Muffle being drawn from the heating pit. The carburizing gas is forced over the work by the fan in the top of the muffle. See Fig. 9 for the method of charging.

One way of making practically the whole furnace enclosure a uniform and diffuse source of radiation is to use electric resistors, for example, of resistor ribbon or of coiled wire, looped over refractory supports set in the walls. Resistors can be so disposed that the walls are quite uniformly heated, and by closer spacing in the cold corners and more open spacing in the hotter zones, as well as by mounting them on the inside of the doors and beneath the hearth, the ideal of a uniformly heated enclosure can be closely approached.

The electric resistor, since it can be distributed over the furnace walls, can release the desired amount of heat per unit of time at a lower temperature than can a flame; it is a more diffuse source. Thus the resistor temperature in a hardening furnace will be of the order of 1800°; a gas or oil flame has a flame temperature approaching 3000°. Overheating by direct radiation, or by letting the flame itself strike the work, may therefore be avoided by the use of resistors.

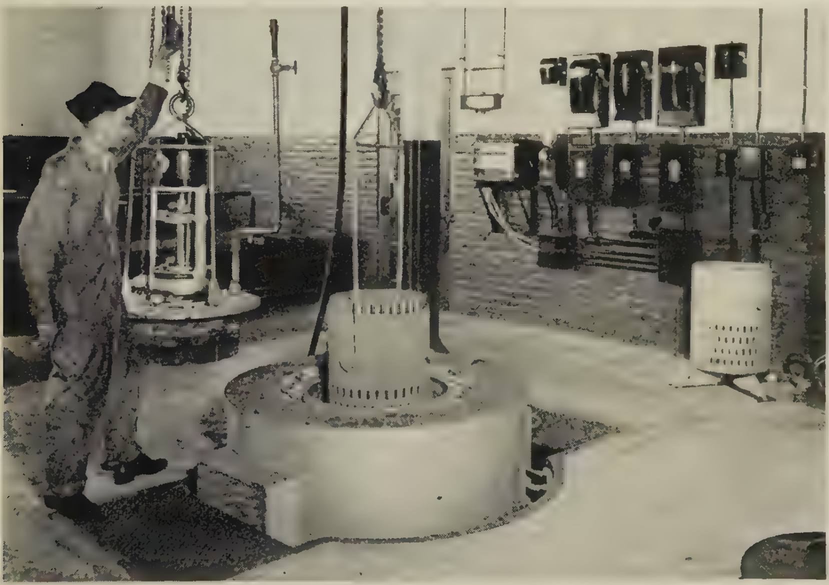


FIG. 9.—The stock to be carburized is charged into a perforated pot so that the gas may be circulated. (Courtesy Hevi Duty Electric Co.)

The diffuseness of the source in, and the ease of, automatic control of resistor furnaces, coupled with the freedom from products of combustion and the ability to use a controlled atmosphere, made the electric furnace closely approach the ideal; and during the past decade well-engineered resistor furnaces reached a high state of development and acceptance.

Controllability of Modern Furnaces. The competition afforded by well-designed electric furnaces to inferior designs fired with gas or oil spurred dealers with those fuels to sell, and manufacturers designing furnaces for those fuels, to improve furnace design so that temperature uniformity could be better approached and to devise ingenious and reliable automatic controls for fuel and air so that temperature control might be as feasible in fuel-fired as in electric furnaces. In general the control devices for fuel were more difficult to design so as to be reliable and trouble-free than were the electric ones, but they have nevertheless been perfected so that either control at one temperature, or "program control" which automatically produces a given time-temperature cycle, can be had either with electricity or with fuel, though it is still probably somewhat questionable whether oil firing is quite as satisfactory from the control point of view as is gas firing. Mawhinney [7] has discussed control devices.

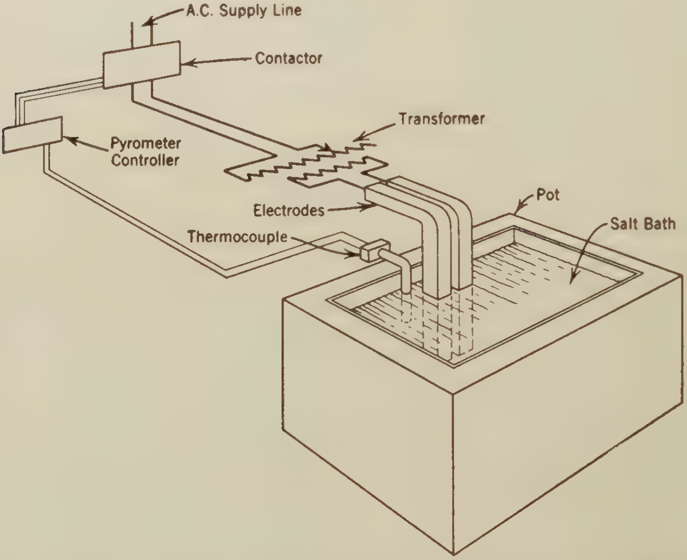


FIG. 10.—Diagram of electrically heated salt bath furnace. The electrodes dip into the fused salt. The hot salt rises and cooler salt is drawn in to take its place and circulation is set up in such a way as to stir the whole bath and distribute the heat in it. Heating of the stock immersed in the salt is thus more uniform than in a stagnant salt bath heated from the outside. (After Hultgren.[21])

In some cases control is better with gas than with electricity because, with the latter, the control is usually on and off; whereas, with properly designed gas furnaces the control is from minimum to maximum. The result is that there is no difference in heating due to difference in radiation, as when the power is off the resistors. Furthermore, the atmosphere may be maintained substantially uniform with minimum and maximum control, which is a condition difficult to effect when the control is on or off. On and off control is much more objectionable with oil-fired furnaces, provided with burners blasting through uncovered openings, than with any other. It is less objectionable when *all* the air passes through the burners.

The difference in control, above noted, may determine the choice between fuel and electricity, for certain operations in which uniform quality and surface are controlling factors. It is, however, entirely feasible to design electric furnaces and controls for minimum to maximum control. For example, this is regular practice in laboratory furnaces for creep testing.

Notwithstanding the equality of control, fuel firing still lagged behind electricity, in that to compete with electric heat for controlled atmosphere furnaces in the treatment of steel, the raw products of combustion had to be kept out of the furnace chamber by a full muffle.

Radiant Tubes. A rather recent development in gas firing made it possible to use controlled atmospheres in gas-heated furnaces without requiring muffles. This is the "radiant tube," a device in which gas * is burned inside a heat-resistant alloy tube, the walls of which radiate heat, while the products of combustion pass on through the tube without entering the furnace itself. The radiant tubes, strung along the furnace walls, suspended in the roof, or lying in the hearth, behave quite the same as though they were huge electric resistors. In order to prevent products of combustion leaking out of cracks or pin holes in the alloy tube, the products of combustion are exhausted from the tube by air-induced draft, by fan, or by an air aspirator, and the tube operates slightly below atmospheric pressure. This method of firing also simplifies the maintenance of an air stream around the central core of burning gas: the air stream protects the tube interior. By avoiding turbulence in the streams of gas and air, combustion may be slowed down so that the flame is very long and free from a zone of excessive temperature. At the point where the flame would otherwise terminate and the tube begin to be cold, the introduction of a "core-buster," to produce turbulence, or the location of a bend in the tube, will allow extending the flame still further. Flames of 60 ft. or more can thus be produced inside a tube, and with reasonably uniform temperature along the tube.[8] The tubes are 4 in. diam. or larger, so that they are analogous to very large resistors only; because of mechanical requirements for support they cannot be located as close together as can resistors; and a radiant-tube furnace interior is not quite the approach to a "sheet of heat" that can be had with resistors. Because of the size of the radiant tubes it is not feasible to make small furnaces to use them; their main application is to big annealing covers for box annealing (see Fig. 5 and Fig. 11; also Fig. 3*b*, Chap. 12) and to large continuous or very large batch-type furnaces. In the annealing cover and analogous uses it is possible to hasten the cooling cycle by passing cold air through the tubes, and in this they have an advantage not possessed by resistors. The hot products of combustion from the tubes may be used recuperatively to preheat the air for combustion.[9]

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* The application of gasified oil or even of pulverized coal firing to radiant tubes appears to be feasible also.

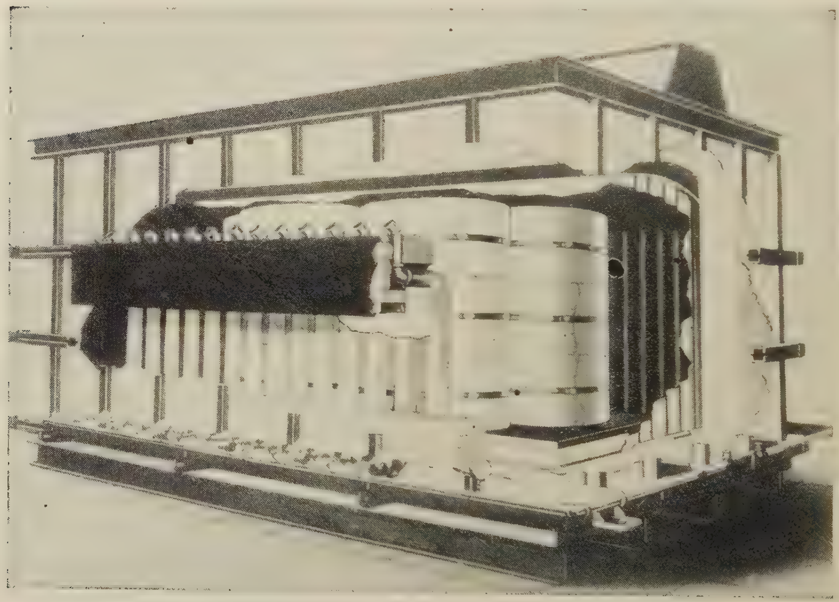


FIG. 11.—Another type of radiant tube annealing cover, with tubes arranged vertically instead of horizontally. (Courtesy Wean Engineering Co.)

Were design and materials worked out for a thin annealing cover or bell that would remain gas tight in service under direct firing of gas into the space between the furnace proper and the cover (and in the cooling stage of the cycle as well), the radiant tubes could be dispensed with.

Limitations of Radiant Tubes. In order to radiate to the charge at the same temperature that a resistor would, or at a somewhat higher temperature because they cannot be made into quite as diffuse a source, the firing face of the radiant tube must be hotter than the radiating face, so for a given furnace temperature the maximum temperature of the metal in a radiant tube must be above that of a resistor; hence the upper temperature limit of usefulness of radiant tubes for satisfactorily long life is below that of metallic resistors. However, this limit is high enough so that the heat-treating operations that require no more than 1750° furnace temperature can be carried out without fear of unduly short life. Some of the applications to annealing of high silicon electric sheet raise a question whether tube life will be economical, and some of the types of radiant tube firing with vertical tubes are thought by some to be more severe on the tubes than is the case with the horizontal system. With experience the safe temperature may be improved, but some applications, like copper brazing in a reducing atmosphere at 1950–2100°, in which NiCr resistors have sometimes lasted five years or more, are not yet within the scope of the radiant tube.

Future Possibilities. The metallic resistors also have their temperature limits, with 2100° as the upper limit for regular NiCr resistors. Using crimped ribbon or tubular resistors, so as to provide the maximum radiating surface per cross section of resistor, and providing some means for circulating the furnace atmosphere so as to carry the heat away from the surface of the resistor as rapidly as possible, tends to allow somewhat higher operating temperature. The tubular resistor elements, mounted vertically, are claimed to set up an automatic circulation by a "chimney effect."[10]

Efforts are being made to produce alloys of higher operating temperatures, and alloys containing, besides Ni, Cr, and Fe, a small percentage of Al, with or without Co, are being commercially introduced that do have higher operating temperatures.[11, 12, 13, 14, 15, 16] These alloys still have drawbacks. Some of them are too brittle to stand cold forming and have to be coiled or bent hot. Some tend to become embrittled with use. Some stretch and sag at operating temperature and require a design that allows proper support. Further development is expected to remedy some of these drawbacks.

Even these alloys will have their upper temperature limit, which will probably fall a bit short of the temperature for hardening of W high-speed steel, though that for Mo high-speed may ultimately be reached.

Molybdenum resistors will stand very high temperatures but cannot be used, bare, in air since they oxidize too readily. They may be used in a hydrogen atmosphere but it is difficult to ensure that in commercial furnaces the hot resistor is never exposed to an oxidizing atmosphere. Molybdenum wires, hermetically sealed into a tightly fitting refractory sheath and provided with terminals for connection outside the furnace, are being developed [17] that may allow commercial operation without necessity for a reducing atmosphere at temperatures even above that of high-speed steel hardening; but their behavior has still to be demonstrated in service.

Non-metallic Resistors. For a temperature range up to that required in forging furnaces, for the present at least, non-metallic resistors are required. Carbon resistors have been commercially used but have limitations which have kept them from cutting much of a figure. Gould and Toole [18] have experimented with means of protecting them for use in ceramic firing, but their feasibility for heat-treating furnaces has yet to be proved. The silicon-carbide (Globar) type may be employed. These "Globar" resistors do not have a very long life at their maximum operating temperature, the life being of the order of 1000 hr., with possibilities of pushing it up to 2000 hr., or more, and with further possibilities in the use of tubular rather than rod resistors. Globar terminals have to be well designed and properly maintained, the resistors are relatively fragile, it is more of a problem to dispose them in the furnace so as to get a diffuse heat source than it is metallic resistors, and only relatively short Globars are available so that as a rule they are suited only for the smaller types of furnaces. Since Globar resistors may change in resistance while they are still usable, a tapped transformer or analogous means for voltage variation is needed.

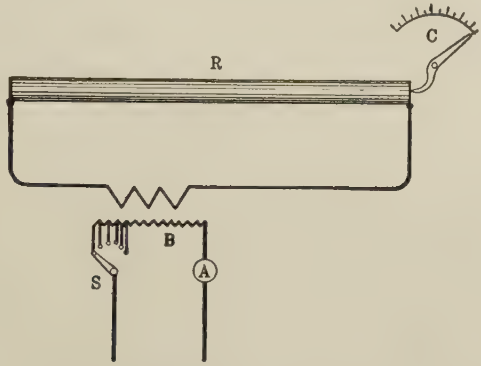
The Globar is one of the few types of heat source commercially available at present that will serve for heating high-speed steel for hardening in a controlled atmosphere free from products of combustion. One could use a silicon carbide muffle externally heated by fuel, or could heat direct with a carbon resistor.

Muffle Furnaces. A simple high-speed heating outfit that offers a useful approach to atmosphere control is a gas-fired furnace with two graphite crucibles serving as muffles, into which the steel is charged, with burners at each end. The waste gases from the high heat are used for preheating. On the next round the other burner is used. Charcoal or carburizing compound in the bottom of the crucible with the work properly supported above it, or a suitable gas atmosphere led into the crucible, will supply a protective atmosphere.

Heating of very large pieces for forging has to be done by fuel.

Neither radiant tubes, metallic resistors, nor Globars are suitable for these high temperatures and large furnaces.

Direct Resistance Heating. So far we have considered only the heating methods by which externally generated heat is brought, by one means or another, to the surface of the piece to be heated. There is another possibility, i.e., making the piece itself carry an electric current and be heated internally by its own resistance. The method is restricted to pieces of uniform cross section and to those whose cross section is small enough so that the currents involved are not too large to handle. It is admirably adapted to tubing and rod, and materials up to 2½ in. diam. in solid rod or the equivalent in tubing are handled. With such a method all problems of holding down the rate of heating fall away, since the whole cross section heats up at the same rate and there is no problem of heat transfer or of shielding of one piece by another. Automatic temperature control can be made very precise. The method is applied in the Snead heat-treating outfit (Figs. 12 and 13) and in the Berwick heater (Fig. 14) which has been chiefly applied to rivet heating and to heating for upsetting or forging (see Figs. 9-11, Chap. 14), though it has very interesting possibilities in heat treatment, especially of objects of uniform cross section.

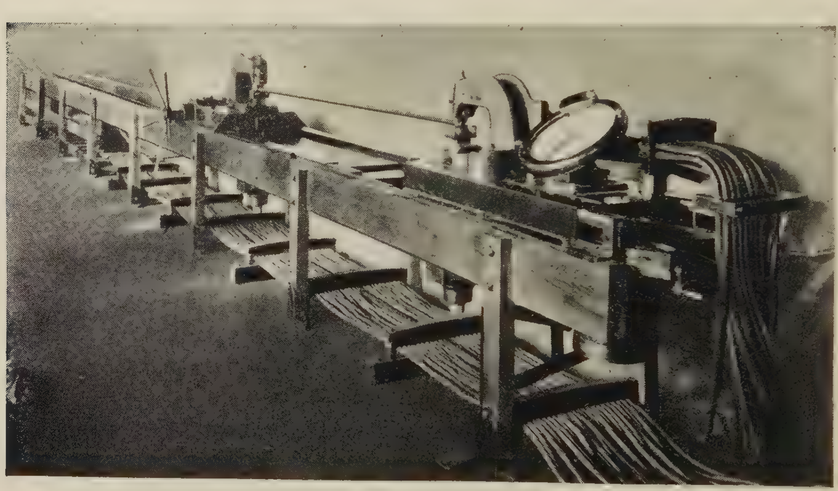


R – Piece Being Heated
B – Transformer
S – Regulating Switch
C – Temperature Indicator Actuated by Expansion of R
Ⓐ – Ammeter

FIG. 12.—Principle of the Snead Process of heat treating.

These resistance methods need careful engineering to avoid cold ends at the contacts, or pitting due to arcing at poor contacts, but they have been developed to a high degree of perfection for specific uses. By applying "electric eye" control, the piece can be brought rapidly to the desired temperature without possibility of overheating or burning, the current cut off, and, if the piece is not removed within the proper temperature range for forging or quenching, as the case may be, the current may be automatically restored.

One interesting application of a Berwick heater is in heat treating high-speed steel for drills. The stock is cut to twice the length of the desired drill and heated between a special heavy copper grip at one end and a mercury cup at the other to keep both shanks soft. The center is heated to the desired quenching temperature, quenched, cut in the center, the spiral flutes ground in, and the tips ground, to make two finished drills. The necessity for stock of uniform cross section limits direct resistance heating to special cases.



(Courtesy Snead & Co., Jersey City, N. J.)

Fig. 13.—The Snead electric heat-treating machine.

Induction Heating. A somewhat analogous method is the heating of stock by its own resistance by means of high-frequency induction. This has been chiefly applied in the "Tocco" process of differential hardening of the bearing surfaces of crankshafts and the like, as mentioned in Chap. 13. By allowing time for the heat to penetrate it could be applied to ordinary heat treatment. This would have somewhat greater flexibility than the direct resistance method, for the high-frequency primary coils could be so wound and spaced as to produce uniform heating on objects of reasonably varying cross section. Specially wound coils would be required for each different size and shape of object and limitations as to irregularity of contour would exist. This method is very speedy and capable of accurate control.

Its application is held back by the first cost of the high-frequency equipment, so that it could be considered only for quantity production. It is being tried in heating for forging.[19]

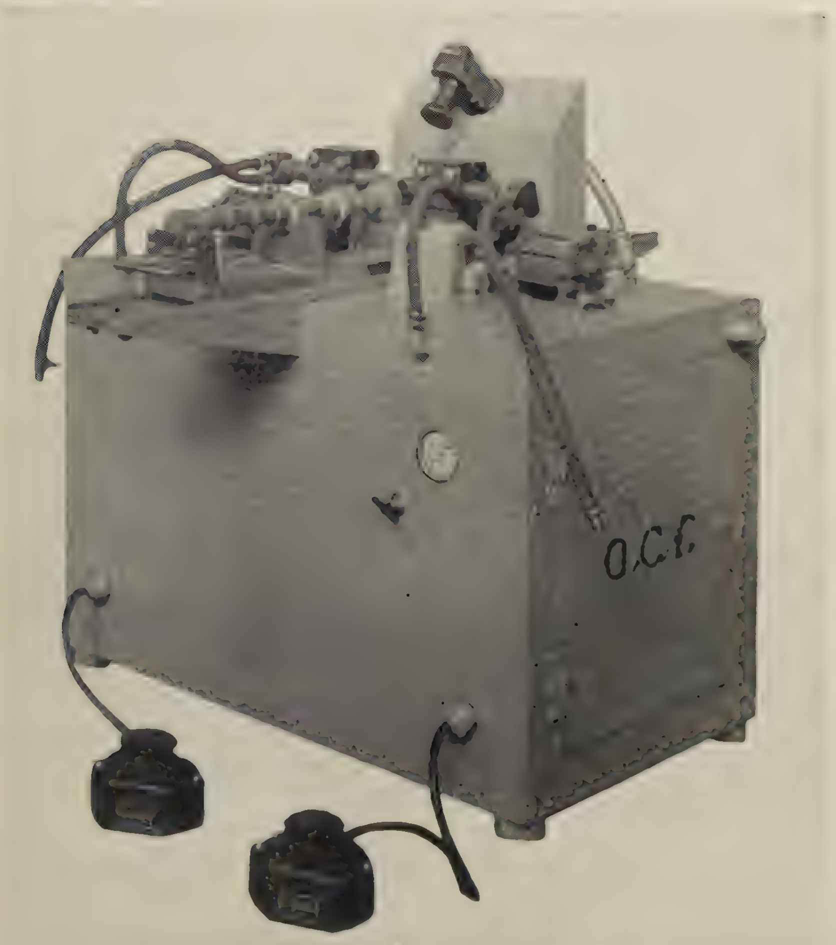


Fig. 14.—Berwick heater used for hardening. The cam tip is placed between the current-carrying contacts and electric current passed through the tip itself. This rapidly heats it to a hardening temperature. The "electric eye" notes when the proper temperature has been reached, turns off the current and turns on the quenching stream. (Courtesy American Car & Foundry Co.)

Choice among Alternative Methods. There is a wide variety of actual and potential heating methods from which the heat treater may choose the one best suited to his particular problem. His choice should be primarily based on the suitability for exact and uniform heating of the particular part to be handled. He should not be misled into accepting a method that will not do a good job by arguments as to the fuel cost of one method compared with another. As Trinks[4] points out, fuel cost seldom amounts to 1% of the final cost of a heat-treated part. It is penny wise and pound foolish to use a furnace with poor temperature distribution or to load a charge so densely as to defeat the possibility of uniform heating, just to save a few cents per day in the cost of fuel. If inspection is close, rejects due to irregularity of heating will overbalance the apparent saving, and if inspection is so lax that variations are not caught, failures in service will cost the user far more in cash and the maker far more in reputation than it would have cost to do uniform heat treatment.

Heat a Means to an End. There is no definite relation between quality and uniformity of heat-treated products and any one form of fuel or electricity or furnace. Each has its limitations and field of usefulness.

Heat treatment is a process of transferring heat to and from the product to be heat treated. It is not merely a process of generating and conserving heat and of controlling temperature. The manner of applying heat to the product is of primary importance.

Heat treating is accomplished with heat—not with fuel or electricity, which are but means to the end. It is a process similar to cooking. The underlying principles apply with equal force in the heat-treating department of the shop and in the kitchen of the ordinary household. The difference is principally in the nature of the product and the "cooking" process.

Fuel and Electricity. The steel doesn't know or care what the source of heat is. It may be affected by the atmosphere that comes from the burning of the fuel and by the direct radiation from a source of too high temperature that overheats the exposed portions, but with the exception of such factors, heat is heat to the steel. It likes uniform heat and dislikes non-uniform heat, but its behavior in non-uniform heating is due to the non-uniformity: it has no likes or dislikes for oil, gas, or electricity as such.

The heat treater should be as uninfluenced by specious arguments for or against a given type of fuel as is the steel itself. He should view the problem solely on the basis of the effectiveness of the fuel and furnace in producing the desired result, and on that of final, overall cost of production, taking all factors into account. In these, first cost of fuel and equipment are seldom the true determining factors.

The long prevalent controversy over the alleged relative merits of different forms of fuel and electricity, prompted by competitive commercial conditions, is beside the point and clouds the real issue. The relation is constantly changing with the development of improved designs of furnaces.

Combination Heating. Under some conditions, the use of more than one means of heating in a single furnace may be considered. Goss [20] reports low heating cost and effective temperature control in a continuous strip normalizing furnace using radiant tubes to heat to 1400° and resistors in the section of the furnace where the stock is heated from 1400° to 1700°. Kjerrman [22] describes a gas-fired annealing furnace, provided with electric resistors located beneath a heat-resisting alloy hearth, for better temperature uniformity.

Comparisons frequently made with claims for advantage attributed to one or another fuel are, in effect, but a comparison of specific types of furnaces operating under specific conditions. The relation may be changed with different types of furnaces or methods of operation in the same plant at the same time.

Fuel Cost Is only One Factor in Overall Cost. Furnaces designed to capitalize the unique and outstanding advantages of electricity frequently make possible results in quality and cost of product and working conditions for the operators far superior to any form of fuel furnace. It is ridiculous to consider electricity on the basis of energy cost alone in light of such conditions which may make any form of fuel relatively expensive regardless of price.

Development of many improved types of gas and oil furnaces that compare favorably with electric equipment of similar general design in automatic control of temperature, atmosphere, uniformity of product, and cost of installation and operation is a striking illustration of the influence of furnace design on the economics of the problem as a whole.

Industry has too long thought in terms of fuel or electricity and not enough in terms of heat application and finished product. The so-called "fuel problem" will be solved in proportion to the manner in which industry learns to think in terms of better heating and cooling, better furnaces, and their influence on the quality and cost of the finished product.

Effective Furnaces for Heat Treatment. Just as in the selection of fuel, many factors enter the choice of a furnace. Some of these are listed in Fig. 15. Obviously a great many classes of objects that have to be heat treated in quantity production differ sufficiently in their requirements from those of other classes to justify the designing and building of a special furnace to do just that job, its capacity being balanced to the production needs.

Continuous Furnaces. Continuous furnaces, with or without atmosphere control as the case requires, using a wide variety of means for moving the stock through the furnace, are generally built to special order, and though most furnace builders have designs for the most-used sizes and capacities which make those sizes approach a standard, modifications can readily be made to suit a particular problem. With the considerable number of competent firms of furnace makers from whom special designs can be had, there is little reason for failing to fit the furnace to the job.

But the investment for a continuous furnace must be justified by reasonable certainty of a steady call for its product. The interest on an idle piece of special equipment that costs as much as a continuous furnace does, would often counterbalance the higher labor cost of batch operation unless operation is truly continuous. The quality of the product might justify the use of a continuous automatic furnace, even though it is not steadily used, over hand operation, if the type of labor and supervision available is not capable of maintaining uniform conditions. This will be discussed further in Chap. 18.

Some of the various types of furnaces available are diagrammatically shown in Figs. 7-9, Chap. 14. A useful discussion of furnace types is given by Trinks[4] and by Mawhinney,[5] whose books should be consulted for a detailed treatment.

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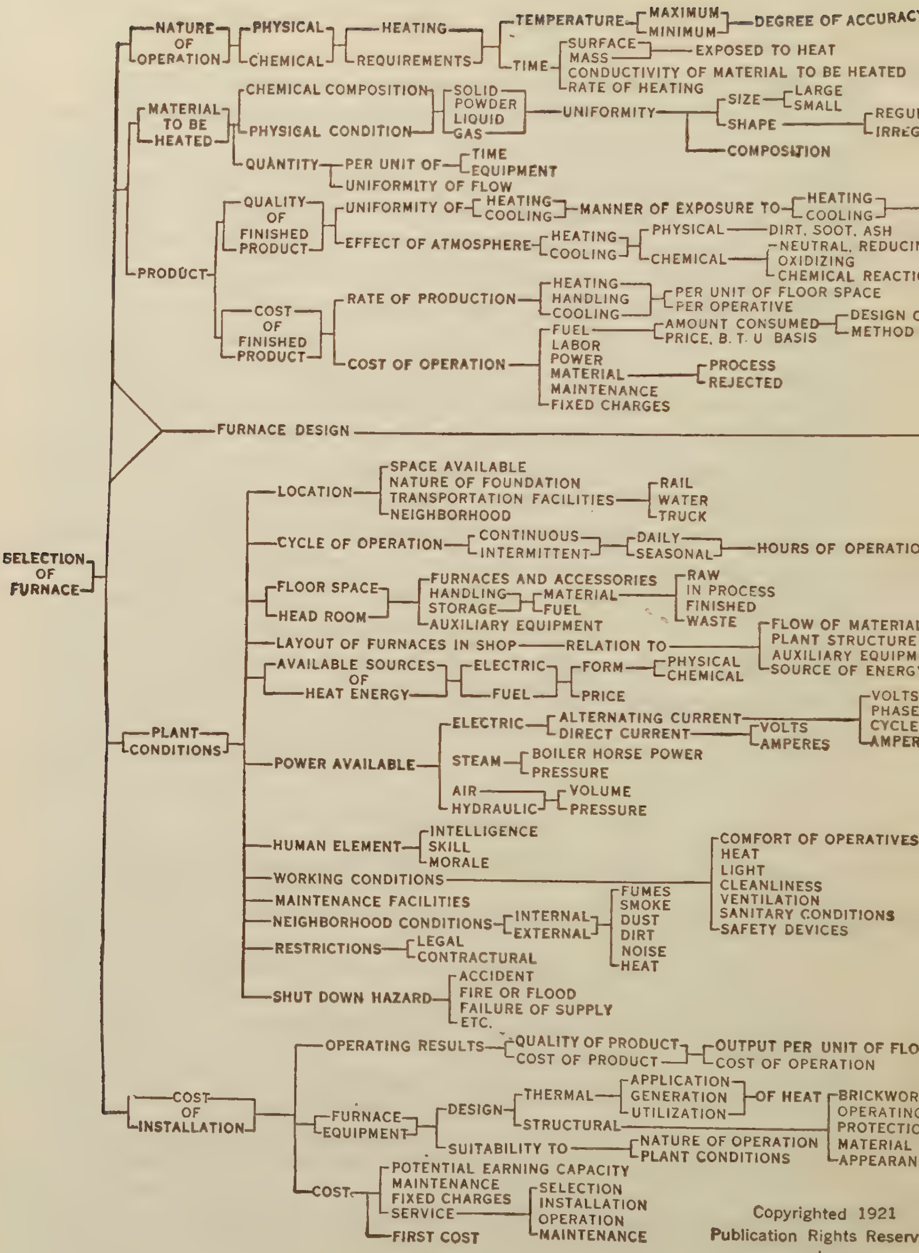

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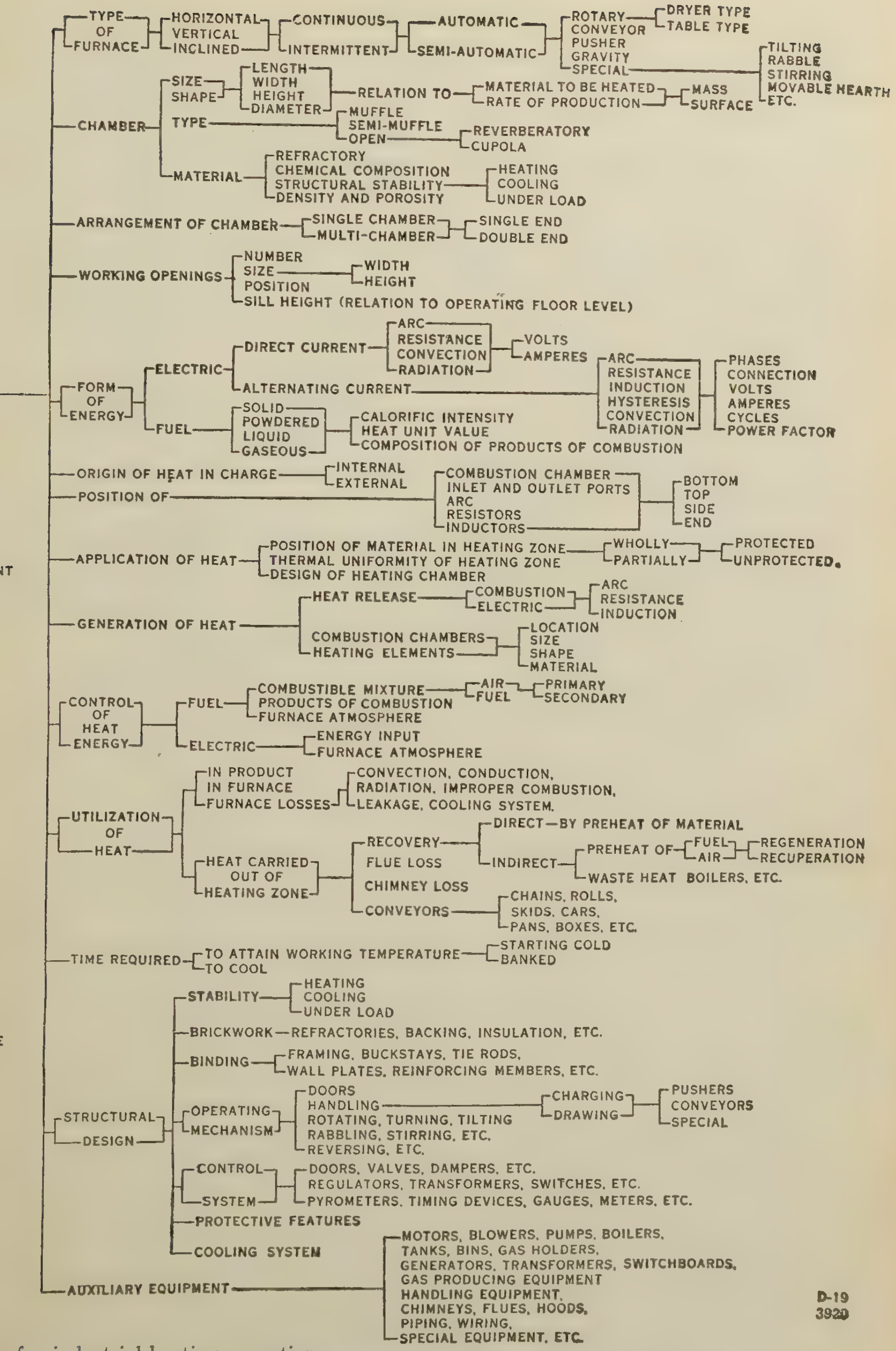
FIG. 15.—Factors governing selectio



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CHAPTER 16

APPLICATION OF HEAT IN HEAT TREATING

Of far greater importance than the kind of fuel or the type of furnace, chosen from among those that can be made to serve, is the way the furnaces are handled in the application of heat to the charge.

Uniformity in Heating and Cooling. Save in the relatively rare cases of differential hardening or local softening, satisfactory heat treatment is predicated upon the piece being *uniformly* heated throughout to the proper temperature, and uniformly cooled throughout at the proper rate. Absolute uniformity through the whole operation can never be achieved, for heat persists in flowing from a region of higher temperature to one of lower temperature and the piece will remain at the same temperature inside and out only when its environment is at that same temperature. Addition or removal of heat is essential and any change in the temperature of the environment produces heat flow to or from the piece.

Heat Flow. The addition or removal of heat must be done through the surface (unless the piece itself is utilized as an electric resistor). The heat flow to or from the interior depends on the rate of temperature change at the surface and on the thermal conductivity and specific heat of the material. If heat will not flow to or from the interior as fast as we wish, because of the resistance of the material to heat flow, there is nothing we can do about it. This is the reason for failure to harden throughout in large sections, and for cracking on heating when heat is applied faster than it can be transmitted through the piece so that thermal expansion or volume changes due to inversions at critical points occur locally instead of simultaneously all through the piece.

The geometrical shape of the piece, as well as its size, influences heat flow. Nothing but a sphere can gain or lose heat at exactly equal rates all over its surface. Even the sphere must be uniformly exposed to the heating or cooling medium, else it will not gain or lose heat in uniform fashion.

When heat treatment is applied to a number of pieces in a furnace charge at once, the mutual influence of the positions of the pieces greatly affects the rate of absorption and loss of heat, so that the problem becomes immensely more complicated.

Arrangement of Charge. The ideal in heating is so to arrange the heat source and the charge that every surface of every piece receives heat at the same rate, and in cooling to so surround each piece with the medium for heat abstraction that all parts of each piece will lose heat at the same rate.

Thermal Gradients. Temperature head is required for heat flow; hence the furnace or heating medium must be hotter than the piece. Actually, in order to heat in a reasonable time, it must be hotter than the final temperature of the piece, else the driving force of temperature difference between inside and outside will be so small that extremely long times will be required to bring the interior to full temperature. Unless the atmosphere is controlled, much damage by scaling or decarburization may be done to the surface during the long holding period.

In practice we are dealing with pieces that have a temperature gradient, with the gradient constantly changing as heating or cooling proceeds. The aim must be so to adjust the heating and cooling conditions that the variations in temperature do not defeat the purpose.

Considering first the heating phase, the piece acquires heat from its surroundings. To heat it uniformly means that the surroundings, the heat source, must all be at the same and, of course, the proper temperature. It is easy enough, with modern control equipment, to hold a given location in a given furnace at a chosen temperature within close limits, as shown by a pyrometer at that point.

Furnace Temperature *vs.* Charge Temperature. It is quite a different thing to ensure that all parts of the furnace are at that same temperature, and still a different one to see to it that all parts of the charge itself are at uniform temperature. The last is what we are trying to do. The manufacture of records from a recording control pyrometer that draw a line showing no change of temperature does not by any means ensure that we are heating other parts of the furnace or all parts of the charge to that same uniform temperature.

Nor does a temperature survey of an empty furnace showing uniformity throughout ensure that a furnace charge will attain that same uniformity. Our aim is to produce uniform heating in the charge; the pyrometer record from the control couple is only a means, not an end, and the end itself can be attained only by intelligent distribution and exposure of the charge. It is fortunate indeed that in most heat-treating operations there is some leeway in temperature, else acceptable results could not be obtained in many furnaces with common methods of loading. The failures that do occur are more often ascribable to deficiencies in temperature distribution in the furnace and to faulty methods of loading than to all other causes.

Temperature surveys of furnaces will often reveal shockingly poor temperature distribution. According to Mawhinney,[1] Fig. 1, taken on commercial furnaces, shows that at normalizing temperature the three furnaces had regions with 50, 75, and 100° variations, respectively, and at process annealing temperatures these rise to 75° in one and about 200° in the others. Such variations have been by no means uncommon in older furnaces; Keeney [2] records cases of poorly designed fuel-fired furnaces with variations up to 250° and unsatisfactory electric furnaces with 60° difference between front and rear.

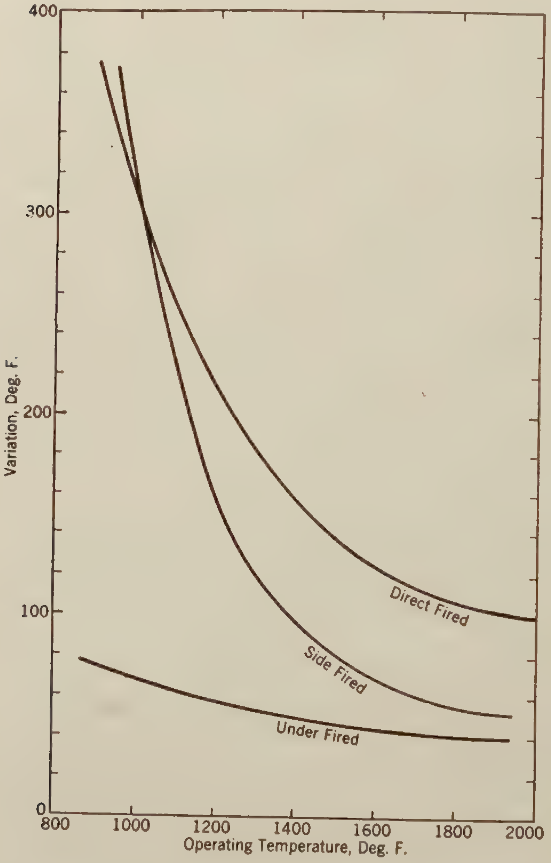

FIG. 1.—Temperature variations in some commercial furnaces. (After Mawhinney.[1])

Uniformity More Difficult to Attain at Low Temperatures. Figure 1 brings out the fact that furnace uniformity is less at the lower temperatures. This is related to the mode of heat transfer.

Heat is transferred by conduction, convection, and radiation. For conduction there must be actual contact of a hotter material with the object to be heated. The hotter material may be, in heat treatment of solid steel, another solid, e.g., a hot hearth, or a fluid, e.g., a salt bath or the gas of the furnace atmosphere. In fluids, convection is concerned as an accompaniment to conduction. The hot fluid is cooled by the piece it is heating, sinks, and is replaced by other hot and lighter fluid, so there is some circulation from this cause, natural convection. Natural convection occurs in a salt bath in an externally heated pot, with the work suspended in the bath.

Forced Convection. Forced convection can be secured by stirring the bath or moving the work. In a furnace, the atmosphere surrounding the work can be made to move; in a furnace where the products of combustion reach the work, by inducing a draft through the furnace by suitable location of burners and exhaust ports. In a controlled-atmosphere furnace using a muffle, electric resistors, or radiant tubes, forced convection can be obtained by circulating the atmosphere by a suitably heat-resistant fan.

The location of the various pieces in the charge is of utmost importance when natural or forced convection is being utilized, since the flow may be easily impeded by close packing, or so facilitated by having channels of low resistance that the fluid follows those channels without bathing the rest of the charge.

This is of extreme importance when a gas is being used as a chemical reagent in gas carburizing or nitriding, since the stagnant places will become depleted of active gas and the channels will receive too rapid a flow. The analogy of rocks in a stream, with stagnant pools, slicks back of rocks, ripples and turbulent flow, and channels, all depending on the distribution of the rocks in respect to the current and the mutual shielding effect of each rock on each other, may be recalled to advantage.

The lower the temperature, the more of the heating is accomplished by conduction and convection. Tempering furnaces are being provided with forced convection, and salt bath furnaces are being heated electrically with electrodes so placed as to produce a pumping, circulatory action. Such means for producing turbulence are highly desirable and may be classed as imperative in the lower temperature ranges.

Forced circulation is being applied even as high as 1300° in bell-type annealing furnaces where each annealing cover is provided with its own fan by which the controlled atmosphere is recirculated. By use of forced circulation in fused salt baths, similar uniformity of convection heating can be had at still higher temperatures. (See Fig. 25, Chap. 12, and Figs. 3–10, Chap. 15.)

The circulation of the heating medium is one of the outstanding accomplishments of recent years. It not only leads to striking results in uniformity but also to notable savings in time and fuel.

Packs and Coils. One of the most difficult problems in securing uniformity of heating and cooling is that of box annealing packs of sheets or coils of wire. A single sheet may be uniformly heated; but, as soon as sheets are stacked, the dead air spaces between them act as effective heat insulators, more effectively the lower the temperature, so that at the relatively low temperature of process annealing heat actually flows to the center of the pack almost entirely by metallic conduction from the edges of the sheets, and very little from the faces of the pack. Forced circulation is of no avail unless the sheets are spaced away from each other by some sort of separator that will allow easy gas flow, and this is ordinarily out of the question because of sagging and warping. As a result, the edges of the sheets and the top of the pack are at temperature long before the center of the pack. It is impossible to maintain a pack or a tight coil of strip or wire with the same time-temperature relation in all parts. Pack annealing is, from the point of view of uniform heating, a metallurgical monstrosity, only put up with through necessity. The space required for individual heating of one sheet at a time for the periods necessary for low-temperature annealing (even though the time could be greatly reduced over that needed in pack annealing) would be prohibitive. Continuous normalizing, with its necessity for only a very brief sojourn at temperature, is being forced by the short comings of pack-annealing methods.

The poor heat distribution in pack annealing is accentuated when very large boxes are used. Some mitigation of the difficulty is obtained by shearing the sheets to the minimum size at which they are usable and providing resistors or radiant tubes between rows of packs in the same box so that the heat flow to the pack is somewhat more uniform. In annealing coiled wire in bell-type electric furnaces it is customary to stack the coils in a frame and put resistor elements within the "hole in the doughnut," as well as outside the bell. No one would think of quenching a coil of wire in a lead bath in a patenting process, for he would know that the interior turns would cool at too slow a rate. (See Figs. 3–7, Chap. 15.)

A little thought about the access of heat and the opportunity for uniform heating or cooling in the close coil at *A* of Fig. 2, compared with the open coil supported as at *D*, will make these points evident. In Fig. 3, note the way the coils of strip are rested on an open spider instead of on a flat base. (See also Fig. 6, Chap. 15.)

When hot-rolled rod is coiled hot, the rate of cooling is very different in different parts of the coil. To secure uniformity in heat-treated wire, it must be heated and cooled as single wire.

Other Furnace Charges. Whereas the difficulties with packs and coils are obvious, analogous troubles arise with the usual furnace charges even of simple shapes, like rods to be mill annealed or batches of steel castings to be normalized. In a car-type heating furnace with a tightly stacked load of castings, the interior of the pile may have been underheated and not refined, the exterior may be overheated, and, when the car is pulled out to cool, the exterior may be normalized to

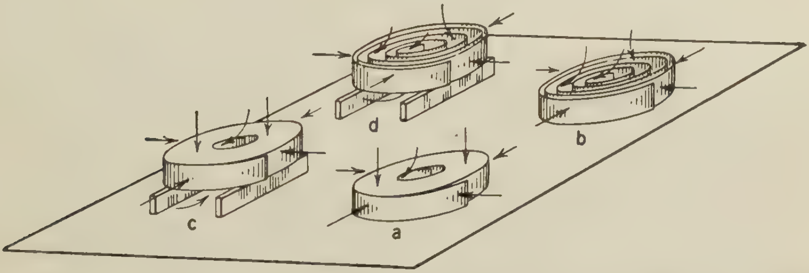

Fig. 2.—Different methods of heating or cooling a coil of metal. (Courtesy W. S. Rockwell Co.)

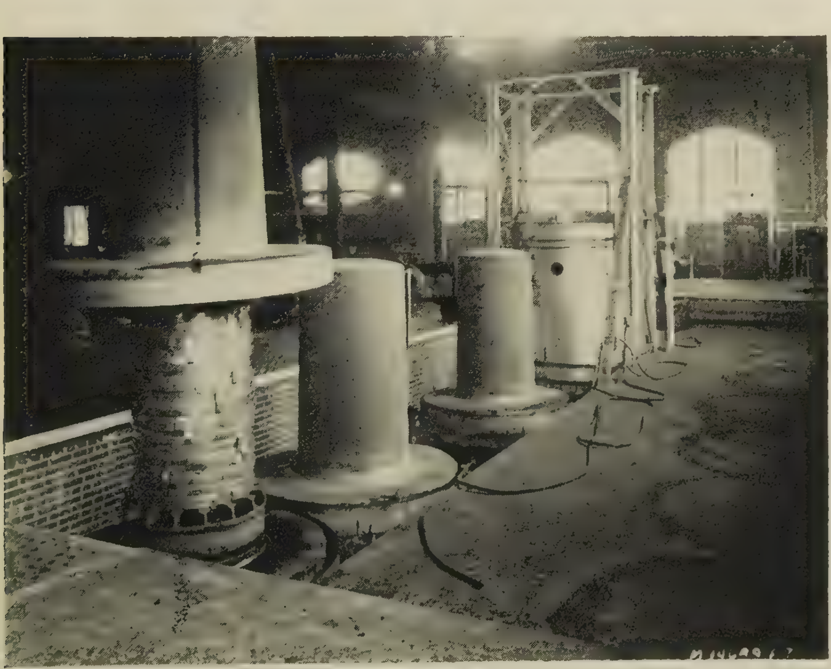

Fig. 3.—Bell annealing of coiled strip. (Courtesy Westinghouse El. & Mfg. Co.)

fine pearlite while the interior cools so slowly that the product is scarcely different from one that is annealed.

People may be misled by glowing advertisements into thinking that, by some magic ability of a specific furnace to overcome the laws of physics, they can heat and cool uniformly with mass methods of loading. It simply cannot be done. The improved uniformity of continuous strip and sheet annealing or normalizing as compared with coils and packs is bringing a fuller realization of this fact.

Peterson [3] makes a specific criticism of mill-annealed bars of S.A.E. 1340, stating that excessive loads are run, so that the time-temperature relations are uneven throughout the charge. He suggests that continuous furnaces should be used, with the charge in one layer, and turned over on its travel, so that each bar may receive the same treatment as its neighbor in both heating and cooling. Similar comments were long ago made by Doyle.[3a]

Modes of Heat Transfer. In conduction and convection heating a tumbling of a charge of small parts or forced circulation of the heating fluid is clearly called for. Convection is principally dependent on the temperature difference between the fluid and the charge being heated or cooled, and on the velocity of flow of the fluid past the surface. It is practically independent of the temperature level at which the operation is being carried on. Radiation, on the other hand, is not only dependent upon the temperature difference, but also upon the temperature level, because it varies as the difference in the fourth power of the absolute temperature. At 2200°, as Trinks [4] points out, one degree temperature difference between the source and the receiver will cause 6½ times as much heat to flow as at 1200°. For this reason, at the higher temperatures of heating for full annealing, normalizing, or quenching, radiation becomes the dominant mode of transfer at the higher end of the heat-treating range.

Directness of Radiation. One outstanding difference between the conduction-convection and the radiation modes of heating is that in the former heat can be transmitted around a corner, while radiated heat, like light, travels only in straight lines. To be heated directly by a radiant source, the charge must "see" that source. If a piece is shaded by another, the piece in the shadow has to get its radiation, not directly from the hot source but indirectly by heat that has been radiated from the hot source to the furnace walls, and caroms around by reflection until it can come in a straight line to the waiting charge. Unlike a billiard ball, hitting a cushion, when heat is radiated to a furnace wall, part of it is absorbed. Not until the inner surface of the wall is heated to a temperature above that of the charge does the heat reflect or re-radiate to the charge. For heating efficiency, a direct path from source to receiver is necessary, but this direct radiation may cause overheating of the directly heated part of the charge when the radiating source is intensely hot, as is the case with flames.

The arrangement of the charge is then very vital in respect to whether it can "see" the hot source, or the furnace walls, or whether it has to get its heat by metallic conduction through its points of contact with parts that are directly heated and through gas conduction and convection. The "umbrella action" of the charge will have far more to do with the temperature uniformity of the charge than the mere temperature distribution in the empty furnace, important as this is.

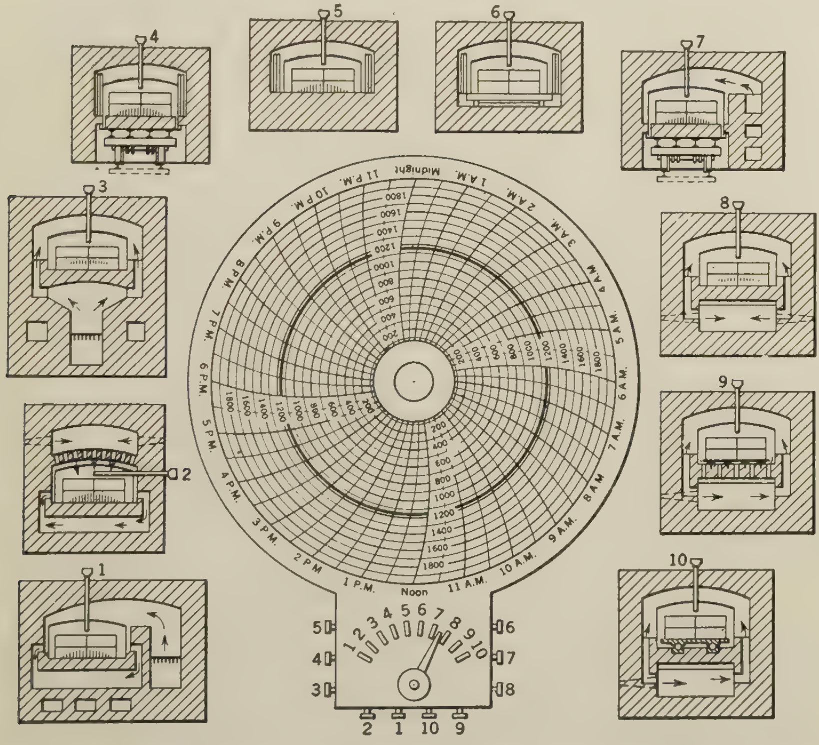

FIG. 4.—The charge may not be uniformly heated even though the pyrometer shows the desired temperature. Shaded portions of the charge indicate regions slow to heat under the loading conditions used.

Exposure of the Charge. The relation of the temperature indicated by the pyrometer to the actual temperature of the product is illustrated by Fig. 4. The pyrometer chart is a reproduction of one of a set, practically identical in character, indicating the temperature control in five furnaces operating on the same twenty-four hour day in the same plant.

Although it might be assumed that a charge would be uniformly heated in any furnace recording such a chart, there would, nevertheless, be a marked variation in the uniformity with which the charge would be heated in some of the different types of furnaces (1 to 10), even though the indicated chamber temperature and the time of heating were the same in each case.

The approximate locations of the incompletely heated zones, reflecting a difference in furnace design or manner of loading, are shown by the shaded sections. The electric furnaces (4, 5, 6) are not exceptions to the rule, for while they may be alike in manner of generating and controlling the input of heat, they differ, like the various fuel-fired furnaces, in the manner of applying the heat to the surface of the individual piece making up the charge in the chamber.

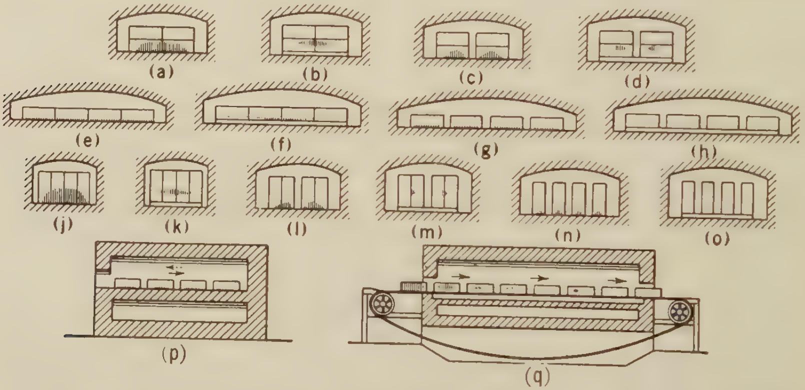

Fig. 5.—Influence of furnace design and manner of loading upon actual temperature of the charge. The indicated temperature of the furnace itself could be the same in all these instances, while the work may be heated non-uniformly. The shaded portions represent locations that heat less rapidly than the rest due to methods of loading which hamper heat transfer.

The product may likewise vary in uniformity in each of these furnaces, depending upon the manner of distributing or exposing the charge (*a* to *q*), as shown in Fig. 5, and with a variation of the mass and surface exposed to the heat, although there may be no variation in indicated chamber temperature or time of heating or cooling.

In considering the relation of the temperature indicated by the pyrometer to the temperature of the product, it is well to bear in mind that the pyrometer merely indicates the temperature at that part of the furnace chamber in which the couple is located; it does not indicate, necessarily, the temperature of the product. Even though the charge is exposed to the heat for a long period without any apparent change in the pyrometer indication, it is unsafe to assume that the pieces in the chamber are heated alike unless they are exposed in the same manner, for the same time, at the same rate, and in the same atmosphere.

The principle involved is similar to that governing the conduct of cooking operations. It is wise to "keep an eye on the biscuits" as well as on the thermometer indicating the temperature maintained at some point in the oven of the stove, or the chamber of the furnace. Cooking on a large scale with automatic equipment, as in the manufacture of bread or crackers, removes many of these variable factors because the product is uniform in character and subject to heat in a manner that provides for individual exposure and control of the factors governing the time and rate of heating or cooling.

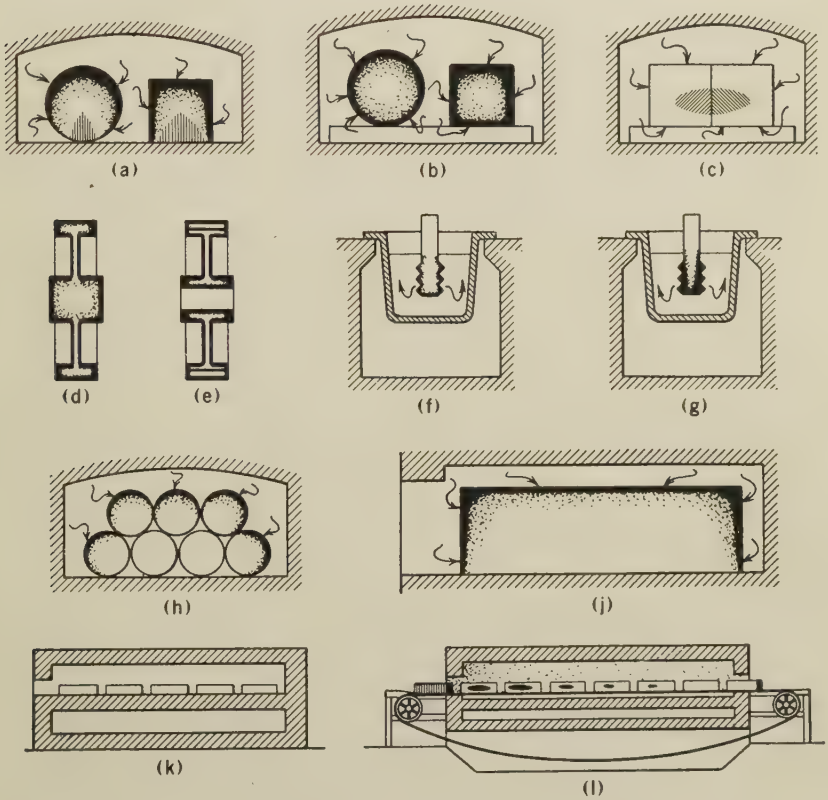

FIG. 6.—Factors influencing time and rate of heating.

Those who are inclined to accept the pyrometer record as evidence of the temperature in the product, without sufficient thought of the variable factors previously referred to, would do well to consider the possibility of producing uniform loaves of bread or crackers on a large scale with the methods of heating, cooling, loading, and exposure so common to the metallurgical industries.

The final test of food, in the case referred to, is the quality of the finished product. The use of pyrometers and control devices is but a means to this end. The same should be true in the heat-treating plant, because heat treating, after all, is but cooking on a large scale.

Control. In some shops we find elaborate provision for indicating and controlling temperature, but little thought of controlling the many other factors affecting the uniformity with which the product is heated or cooled. In many instances, the operation is really one of manufacturing pyrometer records instead of producing uniform metal products.

The economic test of the operation is proof in the form of uniformly heat-treated products, not merely evidence in the form of pyrometer records indicating the temperature at which it may be alleged the product was heated or cooled. Either the product is uniform or it is not. If not, then the testimony of the pyrometer chart is outweighed by that of the testing machine.

Unwarranted emphasis on temperature control as the essential element in heating or cooling operations and on output, fuel consumption, etc., as the determining factors in cost, frequently leads to neglect of the influence of the time, manner of exposure, and the rate of heating or cooling upon the quality and cost of the finished product. It is the acceptance of heat by the charge, not the mere development of heat in the furnace, that is the ultimate goal.

Control of time and rate of heating or cooling is just as essential as control of temperature. Time and temperature are so inseparably linked that it would be well to associate the two as one factor—*"time-temperature."* The use of such a term may tend to discourage the usual abstract consideration of temperature without regard to the time necessary to attain a given temperature and to encourage the thought that there are a number of factors, subject to control, in addition to temperature, that affect the uniformity with which the product is heated or cooled.

The manner in which heat is applied to the surface of the product to be heated, which very largely reflects the design and method of loading and operating the furnace, is of paramount importance and necessarily involves factors, affecting the manner and time of exposure and the rate of heating and cooling, which are as important as the factors that influence control of temperature. All these factors must be considered in order to obtain a product as uniform as the pyrometer indicates it should be.

Methods of Loading. The influence, upon the time and rate of heating, of the exposure resulting from a given manner of loading and from variations in the surface, shape, and section are illustrated by the diagrams in Fig. 6.

A round billet of given mass and weight, placed on the hearth of a furnace (*a*), will heat less uniformly and at a different rate than if placed above the floor (*b*) to permit circulation underneath and uniform application of heat to the entire surface.

Rate of Heating. A billet of the same mass and weight in rectangular form will heat at a different rate and to a different degree of uniformity if placed in either position, on account of the relatively larger surface in proportion to the mass at the corners, which are the first to heat and the first to cool. Even though the shape of the rectangular billet may permit a higher rate of absorption, it is obvious that the difference in shape may actually call for a lower rate of heat transfer to permit a slow rate of heating, in order to prevent a material difference in temperature, or time of exposure to that temperature, between the corners and center of the mass.

The same billet in the form of an irregularly shaped crankshaft or axle with variable sections will heat in a still different manner if placed in either position, because of the difference in surface and section.

A gear blank (*d*) will absorb heat at an entirely different rate from the finished gear (*e*), because of the difference in section incident to the shape of the teeth and the opening for the shaft keyway, etc.

A similar condition may result from the relation of the pieces to each other, even in a chamber that shows every indication of uniform temperature. Thus, two rectangular pieces placed above the floor, with provision for heat application to all surfaces, will heat at a different rate and to a different degree of uniformity when placed in contact with each other (*c*).

Heating in baths of molten metal, salts, etc., is generally considered to be an ideal method for operations in which uniformity of heating and temperature control are essential factors. However, the difference between heating in a molten bath and in a "bath" of gases, with the same method of exposure, in the chamber of a furnace is not as great as is generally supposed. The essential difference is in the influence of the bath upon the surface of the piece, the resistance to sudden changes of temperature, and the conductivity of the bath.

The necessity for suspending the piece in a bath naturally results in an ideal condition for uniform "heat application" to the surface and eliminates many chances for error likely to result from improper methods of loading in the chamber of a furnace, particularly if it is desired to localize the heat, as in hardening the cutting section of a tool.

Inequalities of temperature may exist throughout a bath without being indicated by the pyrometer in its customary fixed position, as may be observed in heating the bath from a cold condition, in introducing a comparatively large mass of cold material, or when the input of cold material does not correspond with the withdrawal of heated material.

Effect of Time. The use of a bath does not eliminate the necessity for considering the time element with reference to the factors that influence the rate of heat transfer and uniformity of heating or cooling. A difference in relation of mass and surface, due to a difference in shape or section, naturally affects the rate of heat transfer and the time required for saturation. Thus, if a tap be suspended in a bath of molten metal *(f)*, Fig. 6, the comparatively thin sections will reach the temperature of the bath before the center of the piece; and even though the center ultimately attains the same temperature *(g)*, the fact remains that the thin sections have been subjected to that temperature for a longer time, which of itself is frequently sufficient to create a difference in structure.

This illustrates the difference between uniform temperature in a furnace and uniformly heated product, and the necessity for considering the influence of the factors that affect time or rate of heating, as well as those that affect control of temperature. Uniform heating is a relative term. Absolute uniformity is rarely, if ever, attained.

The necessity for uniform time and method of exposure and rate of heating or cooling, to produce uniform product, indicates the need for a method of heating, cooling and handling that will provide the same treatment of each piece.

Continuous Furnaces. "Batch heating" or cooling rarely results in uniform product, regardless of the indication of uniform chamber temperature, because of the lack of uniformity in method and time of exposure and rate of heating or cooling. Variation is generally disclosed by the structural difference between the pieces at the outside of the charge and those at the center or bottom.

The advantages of the continuous furnace *(l)* in providing a regular input and output of material and control of the factors affecting time and rate of heating suggest the application of the principle of individual treatment whenever the manufacturing requirements and plant conditions will permit.

Heat Application. The difference between control of temperature in a furnace and control of the application of heat to the product is not merely a function of controlling the input of heating energy, whether it be fuel or electricity. The manner in which the heat is applied to the surface of the piece, which very largely reflects the design and method of operating the furnace, is of paramount importance, and necessarily involves factors affecting the element of time and rate of heating and cooling, which are equally as important as the factors that influence control of temperature.

The ideal condition for the production of uniformly heated product comes nearest to being attained when the heating or cooling medium is uniformly applied to the entire surface of each section or piece and to each piece individually in the same manner, at the same temperature, at the same rate, *for the same time,* in the same atmosphere, in equipment properly adapted to the nature of the process, form of fuel or electricity employed, manufacturing requirements, and plant conditions.

Factors in Heating. Temperature is a more or less fixed factor, determined by the nature of the process and the physical or chemical requirements of the product to be heated or cooled. It should be considered with reference to the distinction between the temperature of the chamber or bath in which a piece is exposed and the temperature throughout the piece itself, which is naturally affected by the time of exposure and the rate of heat absorption.

Time of heating is a variable factor and is influenced by the manner of transferring heat to or from the surface of the piece; the manner of loading, or exposure; the relation of one piece to another and to the source of heat; the difference between the temperature of the piece and that of the chamber or bath in which it is exposed; the rate of circulation over the surface; the conductivity of the material of which the piece is composed; the relation of the mass of the charge to the chamber or bath; and other factors outlined in Fig. 7.

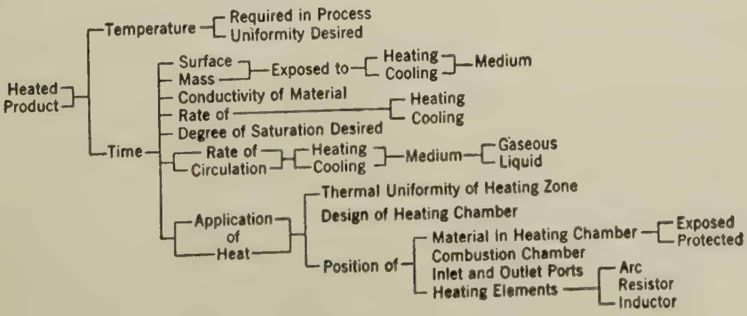

Fig. 7.—Temperature-time factors.

Rate of heating or cooling material must be determined with reference to the physical or chemical requirements of the finished product, and with reference to the mass, shape, section, exposed surface, and manner of exposure of the material, all of which affect the rate at which it *should* be heated or cooled, regardless of how fast heat *could* be transferred.

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CHAPTER 17

CONTROL OF THE COOLING PHASE

Chapters 5 and 6 on normalizing and quenching, as well as comments in several other chapters, have brought out the fact that the rate of cooling is no less vital than the temperature and time of heating.

Too often the importance of the cooling rate is lost sight of and the cooling operation is thought of merely as a means for getting the steel cool enough to handle, forgetting its metallurgical effects.

Mass Cooling *vs.* Individual Cooling. If, in normalizing, the heated charge is drawn from the furnace and dumped on the floor in a mass, or if a closely packed car of steel castings is drawn from the furnace without separation of the castings to allow uniform access of air, the pieces on the outside of the pile may be normalized, but those in the center will not be. In this case air may be looked upon as a quenching medium and its uniformity of temperature and of access to the work, so that the work may cool uniformly, should be guarded almost as carefully as in a liquid quench.

It has been brought out in Chap. 6 that the quenching power of oil and water varies with their temperature and that holding the quenching medium at a definite temperature by cooling coils or the use of a large volume is essential to uniform results. Circulation of the quenching medium or movement of the piece being quenched in such fashion as to avoid the retention of an insulating gas film, or of gas bubbles that retard heat transfer and cause soft spots, is a prerequisite.

Dumping a charge into a quenching bath, instead of quenching the pieces individually or so scatteredly that no piece interferes with the cooling rate of its neighbors, can do even more harm than the nonuniform heating that has been warned against in the preceding chapter.

It is just as necessary to cool individually as to heat individually. The fact that quenching may be a matter of a few seconds, while heating may be a matter of many minutes, or even an hour or more, indicates the necessity for considering the influence of "physical factors" in cooling as well as in heating. Just as in heating, all the factors of time, temperatures, mass, and surface are involved in cooling, and their proper adjustment is as necessary as in heating, in order that the finished product shall be uniformly good.

Necessity for Uniformity. Lack of uniform quenching is responsible for much more variation in uniformity in the product than is generally sensed. Unfortunately, the pyrometer record of the furnace is sometimes accepted as sufficient evidence of uniformity, but in quenching, as in heating, the temperature of the chamber, or of the bath, is not proof of uniform heating or cooling, the test of which is in the product itself. It is as necessary to learn "how to cool" as "how to heat."

Even with stationary electric or gas furnaces used for heating small parts, a quenching machine of some kind is frequently justified even though the furnace itself is properly non-continuous in output.

To avoid warping in quenching of an automobile axle shaft, it may be placed upon rotating rollers that cause it to spin rapidly, thus exposing it uniformly to the stream of quenching liquid that impinges upon it.

Actual Temperatures. The temperature at which the heated product is *actually* quenched is very important in hardening operations. It is not sufficient to heat the product to the desired temperature; *it must be cooled from the desired temperature.* For this reason it is necessary to guard against a drop in temperature when the product is moved to the quenching bath from the heating zone in which it has been heated, or in which it has had a preliminary cooling in case it is desired to quench from a temperature below the heating temperature.

The Human Element. Except under conditions where continuous methods of heating and cooling are in use, this feature becomes more than ever dependent upon the operator, and there is practically no form of automatic control for regulating this phase of the cooling cycle. At this point the operator can either make or break the cycle of operations and lead to success or failure.

Heat treating begins with the initial heating for forming, forging, or rolling, and ends with the final *cooling* phase of the process. The importance of the cooling phase is not measured by the length of this chapter!

CHAPTER 18

THE HUMAN ELEMENT

Batch *vs.* Continuous Operation. In a batch-heating process an individual operator controls the time and temperature, the placing of the charge, and the way it is cooled upon withdrawal from the furnace. He may vary these factors so as to compensate for variations in the stock, as was done in the series recorded on page 199, Chap. 7.

In a continuous furnace operating on quantity production, all these factors are worked out ahead; the furnace controls are set for the desired temperature, the conveyor speed is adjusted, the charge is placed on the conveyor in the same fashion each time, in a manner found by previous experiment to allow suitable uniformity of heating, and the quenching or other cooling is done in a fixed fashion. With such equipment the problem resolves itself into avoidance of mechanical failure of the equipment and vigilant pyrometer checking to see that calibration is maintained. With the equipment operating the same, the time-temperature relation is fixed, and with the same stock, in the same form and dimensions, the result should be uniform, provided the stock is uniform. To secure this last, great pains have to be taken to be sure that the composition and the grain-coarsening and allied characteristics of the stock do not vary.

Adjustment Required Even in Continuous Operation. If they do vary, so that the time-temperature factors require to be altered to compensate, great care must be taken in segregation of the lots that require different treatment. When the operator knows that lot *A* is being followed by lot *B*, which takes a different treatment, he must clear the furnace of lot *A* before changing the furnace conditions and starting lot *B* through.

The operator of a batch furnace has a similar task in miniature. Owing to the smaller lots he deals with, the variations in furnace conditions may have to be made frequently; so the responsibility is usually more directly upon him than on the operator of the continuous furnace whose schedule is set by the metallurgical supervisor. The batch-furnace operator may well be charged with some inspection duties, at least in checking specimens of his output for hardness; so he may have an immediate check on uniformity of heat treatment.

Intelligent Operators Required. The more intelligently he can handle his equipment and the more promptly he knows when any condition is changing so that he can take corrective measures, the more uniform can the product be made.

Some furnace makers say that with the class of labor normally operating furnaces for large production, it is essential to have "only one button that the operator can push"; if anything is left to his judgment there is no certainty whether his decision will be the correct one or not. In such a case there must needs be skilled supervision close at hand, for no matter how good or how automatic the equipment, unforeseen emergencies will occasionally arise.

It takes an entirely different class of artisan from the button-pusher to operate a general heat-treating department where different steels and varying sizes and shapes of pieces require varied temperatures and skilled quenching. Here nothing will take the place of brains, of experience, of common sense, and of an understanding of the principles of heat treatment.

The metallurgist may draw up instructions for heat treatment for a steel of given composition and grain-coarsening characteristics on the basis of time and temperature in heating and cooling. The instructions may apparently be adhered to in the shop without producing material of uniform properties. The charge may be placed in a furnace which is held to the desired heating program of time and temperature, without the charge having the desired and necessary rate of heating and time of holding.

Variations That May Enter. Loading so that different parts of the charge are unequally exposed to heat and cooling in masses so that the cooling rate is unequal will produce non-uniformity of product. Variation in size and shape of the pieces being heated or cooled, and in methods of charging and discharging, may produce highly variable results from the same steel.

Hence not merely time and temperature alone, but the whole complex time-temperature-mass-surface relation must be adjusted. The adjustment has either to be made by the heat treater himself or embodied in specific instructions to him, drawn up by some one who understands that the purpose of heat treatment is not to hold a piece of steel for so long, in a furnace that it is so hot, and then to quench or cool it, but to produce definite changes in the steel by heating and cooling.

If the geometry of the piece or of the charge as a whole calls for a different schedule from that of another piece or another charge, it must be recognized and allowed for in the "specifications for heat-treatment practice" issued to the shop. This allowance requires human intelligence and the necessity for such allowances makes heat treatment in the shop something of an art, instead of the science that it is in the laboratory. The aim should be to bring the shop standard of practice as nearly as possible to the standard which can be reached in the modern laboratory.

Human Control of Variables. The importance of the "human element" in heat-treatment operations cannot be overestimated, even though the practice in some of the largest shops in the country indicates that it is not considered an important factor. The man at the furnace is, as a rule, the ultimate key to the situation and, at the same time, the weakest link in the chain. He is the cook or chef who puts the metallurgical finish on material prepared at great expense by others, and it is his skill and exercise of judgment that very frequently determines the nature of the result and either makes or breaks the cycle of operation.

The Manner of Heat Application. Development of electric furnaces, automatic temperature-control devices for regulating the flow of electricity and fuel, certain forms of automatic heating and cooling equipment, and improved pyrometer equipment have done much to decrease dependence upon the "human element" in the process of making or generating heat. However, it is the manner of applying and utilizing heat, not the process of making or generating heat, that determines the quality of the heat-treated product.

Science and Practice. The practice of the art of heat treating in some shops is out of line with development of the science of metallurgy from the laboratory point of view. The variations in the practice of the art incident to the "human element" are responsible for many difficulties that are improperly attributed to the steel, the furnace, the metallurgist, and other factors exclusive of the "human element" itself. It is necessary to strengthen this weak link in the chain of heat-treating operations if practice of the art is to keep step with development of metallurgical science and improved designs of heating and cooling equipment.

The notion that automatic furnaces with provision for automatic control of temperature and more or less continuous movement of material will, of themselves, eliminate the "human element" is unsound except in more or less rare cases in which there is no material variation in the size, shape, composition, rate of flow or method of heating and cooling the product to be heat treated, and provision to prevent overloading. Such variations make it necessary to consider the influence of the manner of loading and the time and rate of heating or cooling if the product is to be heated or cooled to the desired temperature, and in a manner dictated by the physical condition of the product or by metallurgical requirements. Too often we find modern installations of automatic types of fuel or electric furnaces heating and cooling a great variety of miscellaneous pieces of different size, shape, and weight with little or no thought of the necessity for varying the time or rate of heating by adjusting the load and rate of flow to the requirements. Such conditions frequently have been found responsible for variations in the quality of the ultimate product, as disclosed by the testing machine or in service, in spite of evidence in the form of pyrometer records to the contrary.

Heat Treatment Is an Art. Heat treating should properly be considered as an art or trade that includes not only the process of making or generating heat but also of "cooking" the material to be heat treated with proper methods of heating and cooling to improve quality, and equally proper methods of handling the product and utilizing the heat to reduce the cost.

If, as is generally conceded, men should be paid in proportion to their skill and the part this skill plays in the make-up of the finished product, then a good annealer is as important and worth as much as a roller; and a good man in charge of heat-treatment work is as important and worth as much as a high-grade machine operator or tool maker. In one instance, the man operates a machine, and it is the machine that more or less determines the result after it is set, or, at any rate, acts as a mechanical check on the operation. In the other instance, except when the manufacturing conditions are such as will permit an automatic furnace to eliminate the "human element," there is involved the matter of skill, judgment, and experience on the part of the furnace operator. The furnace and auxiliary appliances are but tools and, while it is desirable that they be of the best, their part in the operation is frequently little more than that of tools in the hands of the operator.

Experienced Men Needed. The "human element" in forging and heat-treatment operations is a controlling factor and the weakest link in present-day practice. The best men available are none too good for such important work. It is necessary that a better type of operator be developed and trained to understand and apply the principles governing the application of heat as well as the process of making or generating the heat. Development of electric furnaces did much to improve conditions. Similar development of fuel-fired furnaces has followed to meet conditions under which it is desirable to employ equipment heated with fuel. Development of known possibilities in heat-treating equipment is retarded by knowledge of the fact that, under present conditions, the manufacturer of such equipment virtually places himself in the hands of inexperienced operators who, all too frequently, have the power to decide whether or not the equipment is suited to the requirements. This particularly is true with fuel-fired equipment, in the operation of which the process of "burning fuel" is frequently given more consideration than the process of heating and cooling the material to be heat treated.

As was stated in the preface to the first edition, it is "the cost of a unit of quantity of given quality" that the heat treater should be interested in, not the cost of one of poor or of variable quality. A "penny wise and pound foolish" policy is never good, whether it is applied to fuel or to the caliber of the workmen.

The results sought in heat-treating practice are determined by the proper combination of the man, the furnace and the steel—not by any one or two of them. Good steel is entitled to proper treatment in good furnaces, and both are entitled to the services of good men in order to produce better results.

INDEX

A

Abnormal steel in carburizing, 219, 266, 281
Ac, 81
Ac_1, importance of exceeding, for grain refinement, 96
Age hardening, 203
Aging, cast steel, 203
  quench, 206
  strain, 207
Air cooling (normalizing), 105
Air-gas ratio in preparing controlled atmospheres, 325
Alloy steel, definition of, 2
Alloys, effect of, on critical cooling rates, 161
Alpha iron (ferrite), 86
  body-centered structure, 86
Alumina in steel, effect in carburizing, 266
Aluminum, additions for grain size control, 220, 236
Ammonia, cracked, as protective atmosphere, 299
  dissociation of, in nitriding, 354, 357, 362
  for nitriding steel, 351
Annealed steel vs. quenched and toughened, 198
Annealing, 94
  atmospheres for, 298
  box, *see* Box annealing.
  bright, *see* Bright annealing.
  C tool steel, 124
  cold-worked steel, 93
  cooling from, 102
  cooling rate after, 115
  covers, 402
  decarburization in, 120
  double treatment, 112
  effect of, on structure, 95
  for maximum ductility, 114
  furnace cooling in, 104
Annealing, heat application, 94
  medium-C steel, 98
  of packs and coils, 416
  relation of hardening to, 150
  semi-hard (0.55-0.75% C) steel, 120
  steel castings, 100
  strip and sheet, 134
  summary of methods, 122
  temperature, 89, 100
  vs. normalizing, 141
  vs. toughening, 198
Ar_1, 81
  depression of, 89
Ar_2, 89
Ar′ (Ar prime), 155, 156
Ar″ (Ar second), 156
Armor plate, carburizing of, 275
Arrests in heating and cooling curves, 81, 88
Atmospheres, circulation of, in furnaces, 415
  for annealing, 298
  for heating high-speed steel, 293
  for normalizing, 301
Atmospheres, controlled, *see* Controlled atmospheres.
Austempering, carburized steels, 288
  definition of, 143
  temperature for, 144
Austenite, cementite absorption by, on heating, 88
  coarse-grained, sluggishness, 237
  coarsening of, on heating, 223, 227
  coarsening temperature of, 231
  definition of, 82
  ferrite absorption by, on heating, 84
  retained, effect on dimensional changes, 179
    in carburized steel, 285
  solubility of C in, 82
Automatic control, 432
Automatic furnaces, 431
Auto-frettage, 183
  Mo-V and Ni-Cr-Mo steels for, 201

B

Banded structures, elimination of, 113
Batch furnaces, 410, 424
Baths, *see* individual types, e.g., Salt baths.
Bend tests, 29, 55
  on austempered stock, 144
Beta iron, 86
Blueing, 298
Borings, cast iron, packing in, for protection of surface, 332
Box annealing, 121
  atmospheres for, 298
Bright annealing, 301
Brinell hardness testing, 8
  table of impressions, diameters vs. Brinell numbers, 10
Brinell strength vs. endurance limit, 37
Brinell vs. tensile, Fig. 8, 21
Brittleness, from grain growth after critical strain, 121
  in impact tests, 44
  Stead's, 121
Burning of steel in forging, 386
Butane for gas carburizing, 273

C

Calcium cyanide in salt baths for cyaniding, 349
Calorizing, 372
Carbide precipitation in Cr steel, 208
Carbon, content of carburized case, factors affecting, 261
  content of carburizing steels, 252
  diffusion of, in cyaniding, 349
  effect on strength, 81
  gradient in carburized case, 259
  range, 5
  solubility in austenite, 82, 252
  vs. toughness, 198
Carbon dioxide, effect in controlled atmospheres, 328, 331
  in gas carburizing, 263
  removal of, 328
  scaling by, 310
Carbon monoxide, in gas carburizing, 263
  inactivity, 308
  production, 321
Carbon monoxide-carbon dioxide ratio in controlled atmospheres, 307
Carburization, mechanism of, 261
Carburized steel, case and core testing, 288
  double quench, 282
  endurance, 290
  hardness of, 284
  heat treatment, 280
  properties, 288
  single quench, 282
  tempering, 284
Carburizers, grain size in, 253
  leather as, 267
  liquid, 348
  regeneration of, 267
Carburizing, 251, 353
  abnormal steel in, 219, 266, 281
  armor plate, 275
  box, oxygen absorption in, 266
    pack, 253, 264
  boxes, 268
    packing of, 269
  compounds, 263, 266
  decarburization in, process, 264
  diffusion in, 257
  effect of alloys on, 255
  effect of Al or Al_2O_3 in steel, 266
  energizers for, 264
  furnaces for, 271
  gas, 271
    packing charge for, 416
  grain size in, 253, 281
  hydrocarbons for gas, 274
    dilution with fuel gas, 273
  mixed gases, 274
  nitriding processes, 353
  normal steel in, 219
  prevention of, 255
  rate of, vs. time and temperature, 255
  S.A.E. steels, 285
  steel, C content, 252
  temperature control in, 270
  tendencies of controlled atmospheres, 328
  use of charcoal in, 265
Carburizing steels, machinability of, 254
Case carburizing, *see* Carburizing.

Case depth, measurement of, 258
  time and temperature, effect of, in cyanide and activated baths, 349
Case testing of carburized articles, 289
Cast-iron borings, 332
Cast steel, aging of, 203
Castability, 58
Castings, Steel, annealing, 100
    Cu-Si, 212
    Cu-Si-Mn, 212
    normalizing, 135
Cementite, absorption of, by austenite on heating, 88
  definition, 74
  hardness of, 156
  properties, 78
  spheroidal, 125
"Chapmanizing," 354
  rate of case formation in, 355
Charcoal, carburizing with, 266
  gas producers, 322
  packing, for protection of surface, 332
Charpy impact testing, 48
  conversion to Izod, 52
Chromium-molybdenum steel for carburizing, 281
Chromium-molybdenum-vanadium steel, secondary hardening in, 215
Chromium-nickel steel for cold treatment, 163
Chromium-vanadium steel for carburizing, 281
Chromizing, 372
Cladding, 371
"Cloudburst" hardness testing, 10
Cobalt in precipitation hardenable alloys, 216
Cold treatment, 161
Cold working, tempering steel after, 199
Compressive strength, 31
Conduction in heating, 414
Continuous furnaces, 409, 424, 429
Contraction in tensile tests, 25
Controlled atmospheres, 292
  air-gas ratio in preparing, 325
  CO-CO_2 ratio in, 307
Controlled atmospheres, carburizing and decarburizing tendencies, 328
  effect on C and alloy steels, 304
  effect of increased methane in, 325
  flue gas as source of, 329
  for heat treating, 292
  for heating high-speed steel, 294
  free oxygen in, 329
  furnaces for, 302
  gases for, 303
    cost of, 322
    drying, 328, 335
  moisture effect in, 311, 317
  oxidizing and reducing atmospheres, 292
Convection, forced, in furnaces, 415
  in heating, 413
Cooking, analogy to heat treatment, 421, 432
Cooling, curves, in various quenching media, Fig. 15, 171
  rates, 152
    after annealing, 102, 115
    after drawing, 196
    air, 131
    importance of control of, 427
    for martensite and pearlite, 156
  slow, effect on structure, 90
  time-temperature relations in, 422, 425
Copper, cast steel for crankshafts, 213
  solubility of, in ferrite, 210
  steel, precipitation hardening in, 209
    surface roughening, 213
      avoided by uses of Ni, 213
Copper-silicon steel castings, 212
Copper-silicon-manganese steel castings, 212
Cored grains, 84
Corrosion fatigue, 43
Corrosion testing, 43
Cracking in quenching, 178
Crackless plasticity, 41
Creep, resistance, 141
  testing, 32
  vs. grain coarsening propensities, 245
Critical cooling rate, 156, 164
  definition, 167
  vs. grain size, 238
  vs. quenching media, Fig. 16, 172

Critical points, 81
  determination of, 88
  merging of, 86
Critical strain, grain growth after, 122
Critical temperatures, 7
"Crystallization" misnomer for fatigue, 33
Cupping tests, 56
Cyanide case, depth of, effect of time and temperature on, 346
  nitrogen in, 345
Cyanide hardening, mechanism of, 346
  of steel surfaces, 345

D

Damage line endurance testing, 41
Damping, 56
Decarburization, 306, 312
  by hydrogen, 311
  by moisture, 297, 311
  effect of, on springs, 296
  in annealing, 120
  in carburizing process, 264
  high-C steel, 332
  tendencies in controlled atmospheres, 328
  vs. scaling, 293
Decarburized surface, effect of, on nitriding, 329, 353
Deep hardening steels, effect of grain size, 237
Diffusion, 150
  in annealing, 107
  in carburizing, 257, 260, 268
  in steel, 106
  of C and N in cyaniding, 349
Dimensional changes in quenching, 177
Directional properties, 93
Distortion in heat treating carburized steel, 282, 289
Double annealing, 112
Draw temperatures, effect of, 195, 197
  range, 193
Drawing, effect of time, 195
  to soften, 192
Dry gas producers, 322
Drying gases for controlled atmospheres, 328, 335
Ductility, 25
  directional, 30
Ductility, effect of toughening temperature on, 194
  engineering requirements, 28
  transverse, 31
  treatment for strength, 114
Dynamic strength, 80
Dynamic tests, 8, 34

E

Elastic limit, 24
Elastic modulus, 27
Electric eye in electric heating for forging, 405
Electric furnaces, 394
Electric heating, 182
  effect on scaling, 298
  for forging, 386
  for surface hardening, 366
  steel for, 368
Elements, symbols for, 3
Elongation, in bend tests, 55
  in tensile test, 25
Embrittlement by cold work, 93
Endurance, 33
  axial vs. rotary beam tests, 35, 43
  carburized articles, 290
  corrosion in, 43
  damage line, testing, 41
  limit, austempered stock, 145
  testing, 35
  vs. Brinell, 36
  vs. tensile strength, 37
Energizers for carburizing, 264, 268
Equilibrium diagram, Fe-C, 81
Erichsen cupping test, 56
Etching, deep, 7
Eutectoid steel, 75

F

Fatigue, *see* Endurance.
Ferrite, absorption of, by austenite on heating, 84
  "beta" modification, 86
  definition of, 74
  hardness of, 156
  properties, 78
  rejection of, on cooling, 102
  solubility of C in, 204

Ferrosilicon in carburizers, 268
File hardness, 14
Flame, luminous, in forging, 378
Flame hardening of surfaces, 364
  steel for, 365
Flue gas, as source of controlled atmospheres, 329
  effect on scaling, 292
Ford crankshaft alloy, 213
Forging, burning of steel in, 386
  electric heating in, 386
  furnaces, 378
    slot heating vs. slugs, 384
  heating for, 375
  overheating in, 376
  supervision of, 407
Forgings, large, normalizing, 138
Fracture tests on hardenability, 232
Frequency curves in statistical methods, 65
Fuels, choice of, 408
  fluid, 391
  for heating, 391
Furnace cooling in annealing, 104
Furnaces, *see also* individual types, e.g., Forging furnaces.
  choice of, 409
  circulation of atmospheres in, 401
  designs, 380
  efficient vs. effective, 392
  for controlled atmospheres, 302
  insulation, 394
  loading for uniformity, of heating, 413
  storage of heat in, 394
  temperature of, vs. charge temperature, 413
Fused baths for heating steel, 342

G

Gamma iron (austenite), 86
  face-centered structure, 86
Gas-air ratio in preparing controlled atmospheres, 325
Gas carburizing, 271
  furnaces for, 271
Gases, dewpoint and humidity, 335
  in metals, 5
  mixed, carburizing with, 274
Gears, distortion in, 289
Generator gas, 323
  adjustment for hardening, 331
  composition, 325
Grain coarsening, 166
  by overheating, 97
  in normalizing, 238
Grain growth, after critical strain, 122
  inhibition of, by solid particles, 236
Grain size, 219
  A.S.T.M. standards, 226
  beyond Ac_3, 90
  in carburizing steels, 253, 281
  coarsening on heating, 90
  control, Al and V additions for, 220, 234
    methods for, 234
  effect in tool steel, 296
  effect of, on deep and shallow hardening steels, 237
    on type of heat treatment, 281
  effect of hot-work on, 91
  effect on impact, 52
  importance of, 246
  in deep drawing steels, 248
  "inherent," 166
    misnomer, 237
  other than A.S.T.M. standards, 228
  value of, 242
  vs. Al additions, 220
  vs. critical cooling rates, 238
  vs. hardenability, 219, 237
  vs. high temperature properties, 245
  vs. impact, 244
  vs. low-temperature properties, 247
  vs. machinability, 240
  vs. mechanical properties, 239, 243
Grain structure, 90
  refinement of, 90
Grains, distortion in cold working or machining, 93
Graphite blocks for protection of surface, 332

H

Hardenability, 8
  factors affecting, 169
  fracture tests for, 232

Hardenability of tool steel, 243
  Shepperd penetration-fracture test, 232
  vs. grain size, 166, 219, 237
Hardenability tests, hardness penetration, 232
  on carburized steel, 232
  surface and center method, 234
  variation in similar steels, 233
Hardening, Age, *see* Age hardening.
  clean, 301
  depth in C steel, 167
  heating for, 147
  on welding, 142
  precipitation, *see* Precipitation hardening.
  principles of, 147
  relation to annealing, 150
  secondary, *see* Secondary hardening.
  surface, *see* Surface hardening.
Hardness, "cloudburst" testing, 10
  definition of, 8
  hot, 11
  Rockwell test, 12
  scales, conversion between, 15
  Shore scleroscope test, 13
  shot-blasting to evaluate, 10
  vs. tensile strength, 20
Hardness conversions, 15
  Brinell-Rockwell B, Fig. 6, 20
  Brinell-Rockwell C, Fig. 2, 16
  Brinell-Vickers, Fig. 3, 17
  Rockwell C-Vickers, Fig. 4, 18
  Rockwell C-Scleroscope, Fig. 5, 19
Hardness testing, 8
  Brinell, 8
  file, 14
  Herbert pendulum, 15
  hot, 11
  mutual indentation, 10
  precautions in, 13
  Rockwell cone test, 12
  scratch, 14
  Shore rebound, 13
  Vickers, 11
Heat, absorption by charge, 423
  application of, 412
  flow in steel, limited rate of, 412
  source, 396
  storage in furnaces, 394
Heat-resistant alloys, effect of controlled atmospheres on, 334
Heat transfer, 396
Heat treatment, 4
  carburized steel, 280
    distortion in, 283
  control of, 375
  definition, 4, 81
  S.A.E. carburizing steels, 286
  supervision of, 430
Heating, for forging and rolling, 375
  for rolling, 387
  fuels for, 391
  internal, 392
  loading of furnaces for uniform, 413
  rate of, 101
  refining structures by, 148
  "saturation" in, for annealing, 95
  steels for low thermal conductivity, 96
  structural changes in, 87
  time of, during annealing, 102
  time-temperature relations in, 422, 425
  uniformity of, 379, 413, 414
Herbert pendulum hardness testing, 15
High-speed steel, atmospheres for heating, 294
  cyaniding and nitriding, 352
  decarburization of, 295
  protecting, by packing during heating, 295
High-temperature service, treatment for, 141
High-temperature testing, 32
Homogenization, 131, 135, 150, 164,
  by heating, 95, 106
Hot hardness testing, 11
Human element, 429
  in control of cooling phase, 428
  in heating for forging, 388
Hydrocarbons, as protective atmosphere, 332
Hydrogen-methane ratio in controlled atmospheres, 306
Hydrogen-water vapor ratio in controlled atmosphere, 311
Hyper-eutectoid steel, definition, 76
Hypo-eutectoid steel, definition, 75
  spheroidizing under Ac_1, 115

I

Impact, on carburized articles, 289
  resistance, effect of temperature on, 52
    of austempered stock, 144
  specimens, effect of dimensions of, 50, 52
    test bar form, 48
  vs. grain size, 244, 247
Impact testing, at low temperature, 50
  Charpy, 48
  conversion of Charpy to Izod, 52
  effect of grain size on, 52
  Izod, 48
  notched bar, 44
  rail steel, 53
  repeated, 47
  service evaluation by, 53
  tensile, 54
  torsion, 54
Inclusions, 6
Induction hardening, for balancing stresses, 180
Induction heating in forging, 406
Insulation of furnaces, 394
Internal heating, 392
Iron, allotropic forms, 86
Izett steel, 208
Izod, conversion to Charpy, 52
  impact tests, 48

K

Kip, 22

L

Lead, alloy baths for tempering, 192
  baths, for heating tool steel, 296, 342
  quenching into, 143, 160
Leather as carburizer, 267
Liquid carburizers, 348
Low-temperature impact tests, 50
Luminous flame in forging, 378

M

Machinability, annealing C tool steel for, 124
  of carburizing steels, 254
  testing, 7, 57
  vs. grain size, 240
Magnetic testing, 55
  magnaflux method for flaws, 55
Manganese steel, grain growth in, 238
  medium, normalizing of, 238
Martensite, alpha, 155
  beta, 155
  cracks in, 143
  definition, 155
  hardness of, 156
  primary, formation, 156
  transformation at constant temperature, 144
McQuaid-Ehn test, 220, 222
  limitations of, 222
Merit indices, 58
Methane-hydrogen ratio in controlled atmospheres, 306
Microscopic examination, 6
Modulus of elasticity, 27
Moisture, effect of, in controlled atmospheres, 311, 317
  produced in controlled atmospheres, effect of, 319
  removal of, 335
Molybdenum, effect on temper brittleness, 206
  in precipitation hardenable alloys, 215
Muffle furnaces, 404
Mutual indentation hardness testing, 11

N

Natural gas for carburizing, 274
Network, cementite, annealing to absorb, 130
  of ferrite or cementite, 90
Nickel-chromium steel, for carburizing, 281
  temper brittleness in, 206
Nickel-molybdenum steel, for carburizing, 281
Nickel-molybdenum-vanadium steel, for turbine forgings, 140
Nickel steel (4.5%), quenching, 183
Nickel steel (5%), effect of cooling rates, 176
Nitrided case, hardness and toughness, 355
  properties, 359

Nitrides, precipitation hardening by, 207
Nitriding, 350
  carburizing processes, 353
  containers for, 360
  effect of alloy content of steel in, 352
  effect of temperature on, 355
  effect on decarburized surface, 329
  growth in, 358
  preparation of surface for, 352
  protection against, 359
  rate of, 355
  steels, precipitation hardenable, 214
  time required, 357
  tool steel, 352
  warping in, 358
Nitrogen, as protective atmosphere, 303
  diffusion of, in cyaniding, 349
"Non-deforming" steel, 179
Normal steel in carburizing, 219
Normalizing, 105, 131
  atmospheres for, 299
  effect of rate of cooling in, on properties, 105
  for structure stability, 141
  forgings, 138
  grain coarsening in, 238
  initial structure for, 106
  prior to carburizing, 254
  sheet and strip, 134
  single pieces vs. a mass, 427
  steel castings, 135
  structures produced by, 105
  temperature uniformity in, vs. pack annealing, 417
  tempering after, 139
  vs. annealing, 141
Notch effect, 38
Notched bar tests, endurance, 40
  impact, 44
  slow bend, 45

O

Oil, for carburizing, 274
  vapor, as protective atmosphere, 332
Olsen cupping test, 56
Overheated steel, refining, 109
Overheating in forging, 376
Overstressing in fatigue, 33
Oxides, precipitation hardening by, 207
Oxidizing atmospheres, 292
Oxygen, absorption in box carburizing, 266
  free, in controlled atmospheres, 329

P

Patenting, 142
Pearlite, C content, 132
  definition of, 75
  effect of annealing, cooling rate on, 115
  emulsified, 133
    as initial structure for spheroidizing, 126
  hardness, 156
  lamellar, 132 (*see also* Troostite, primary).
    variation of impact with fineness, 141
    variation of strength with fineness, 133
  properties, 78, 133
Pendulum hardness, 15
Phosphorus segregation, 6
Plate, carburizing, 275
Precipitation hardenable alloys, of Fe, Mo, W, Co, V, 215
Precipitation hardening, 203
  agglomeration of particles and softening after, 205
  alloys, 215
    solubility in, 205
  by Ni-Al compounds, 213
  in nitriding steels, 214
  of Cu steel, 209
Probability methods in testing, 65
Proof load, 24
Propane for gas carburizing, 273
Proportional limit, 24
Pyrometer readings vs. charge temperature, 413, 419

Q

Quench, regenerative, 281
  single and double, for carburized steel, 282
  toughening, 177, 199.

Quenching, brine, 173
  caustic soda, 173
  circulation in, 170, 175
  cracking and warping in, 178
  dimensional changes in, 177
  effect of surface condition in, 174
  from temperature below that of heating, 182
  hardening, "T-max." in, 165
  in spray or pressure, 176
  mass effect in, 165
  media, 167
    severity of, 182
    temperature of, 175
    vs. toughening temperature, 197
  mild, 182
  oil, 171
    requirements for, 173
  pressure, 176
  single pieces vs. a mass, 428
  size effect in, 165
  stresses, 177, 180
  sulphuric acid, 173
  surface hardening by, 364
  temperature, effect on critical cooling rate, 165, 177
  time, 177
  vapor jacket in, 170
  water, 170

R

Radiant-tube furnaces, 401
Radiation, 397
  in heating, 414, 418
Radiography, 55
Rail-ends for heat treatment, 199
Rail steel, impact, 53
Reducing atmospheres, 292
Reduction of area in tensile tests, 25
Refining, overheated steel, 109, 110
  structure by heating, 148
Refractories, insulating, 395
Resistor alloys, effect of controlled atmospheres on, 334
Resistors, effect of controlled atmospheres on, 400
  for electric furnaces, 399
  nonmetallic, 404
Rockwell cone test, 12
Rockwell C vs. tensile, Fig. 7, 21
Rolling, heating for, 387

S

"S" curves, for pearlite and martensite transformations, 160
Salt baths, activated, for cyaniding, 349
  cyanide in, 344
  for heating steel, 342
  for tempering, 192
  materials for, 343
"Saturation" in heating, 95
Scaling, 378
  by CO_2, 310
  effect of flue gas on, 292
  electric heating, effect on, 298
  in annealing, 120
  prevention of, 292
  "smoke" burners for prevention of, 292
  S in gas, effect on, 335
  time-temperature effects on, 297
  vs. decarburization, 293
Scleroscope hardness test, 13
Scratch tests, 14
Screw stock, cold drawn, tempering. 200
Secondary hardening, cases due to precipitation hardening, 215
  of V steels, 215
Segregation, 6
Shallow hardening steels, effect of grain size, 237
Shearing strength, 31
Sheet steel, annealing and normalizing, 134
Shock testing, 44
Shore hardness, 13
Siliconizing, 372
Size effect in quenching, 165
Slow bend tests, notched bar, 45
"Smoke" burners for prevention of scaling, 293
S-N curve, endurance testing, 36
S.A.E. steels, for carburizing, 285
  heat treatment, 286

S.A.E. 3130, statistical data on, 62
S.A.E. 6130, statistical data on, 59
Sodium cyanide baths for surface hardening, 348
Solid solutions, 82
Sorbite, 186, 192
  definition of, 133
  stages in production of, 187
  structure of, 133
Spark testing, 5
Spheroidization, 132, 141
Spheroidizing, 124
  anneal, 110
  high-C steel, 130
  hypoeutectoid steel under Ac_1, 115
"Split" transformation, 157
Stability of structure, normalizing for, 141
Static strength, 80
Static tests, 8, 34
Statistical methods, use in testing, 59
Stead's brittleness, 121
Steel, definition of, 2, 74
  slowly cooled structure, 76
  structure, 2
Steel castings, *see* Castings, Steel.
Stress concentration, 38
Stress-raisers, 38
Stress, induced, used to resist applied stress, 183
  internal, effect on impact and endurance, 180
  relief, annealing, by tempering, 140 180, 187
    of turbine forgings, 140
    of welds, 142
  strain-diagrams, 25, Fig. 9, 22
Stresses, balancing of, 181
  set up in quenching, 177
Stretcher strains, 208
  in deep drawing, vs. grain size, 248
  stabilization against, 208
Strip steel, normalizing, 134
Structure affected by annealing, 96
Structures characteristic of transformation temperature, 157
Structures produced in heating through the critical range, 148
Sulphur, in gas, effect on scaling, 296, 335
Sulphur, segregation, 6
Surface hardening, by electric induction heating, 366
  by quenching, 364
  grain size in, 369
  Tocco process, 366
Surface reactions in carburizing, 251
Symbols for chemical elements, 3

T

"T-max.," in quench hardening, 165
  relation to grain coarsening, 238
Temper brittleness, 197, 206
  effect of Mo on, 206
  in Ni-Cr steels, 206
Temper colors, 187
  limitations of tempering by, 188
Temperature, control of, 400
  for annealing, 89, 100
  of furnace vs. charge, 413
  gradient in heating steel, 413
Tempering, after normalizing, 139
  air circulation in, 192
  carburized steel, 284
  cold-worked steel, 199
  cooling after, 189
  definition of, 186
  effect of time, 195
  hardening strains, 187
  liquid baths for, 190
  on plates or in sand, 191
  residual stress removed by, 200
Tensile impact testing, 54
Tensile strength, vs. endurance limit, 37
  vs. hardness, 20
  vs. Rockwell C, Fig. 7, 21
Tensile testing, 20
  applications of, 28
  definition of, 20
  dimensions of test bars, 22
  precautions in, 26
  rate of loading, 26
Tensile vs. Brinell, Fig. 8, 21
Testing, *see also* individual tests, e.g., Tensile testing.
  object of, 4
  use of statistical methods in, 59
Thermal analysis, 88

Thermal conductivity, relation to heating rate, 96
"Time" quenching, 177, 199
Time-temperature relations, in heating, 96
  in heating or cooling, 422, 425
"Tocco" process of surface hardening, 366
Tool steel, annealing, 126
  "body" of, 243
  "brittle timbre," 295
  effect of grain size, 296
  hardenability, 243
  nitriding, 352
  nondeforming, 189
  protection of, during heating, 295
Torsion impact testing, 54
Torsional strength, 31
Toughness, definition of, 192
  vs. annealing, 198
  vs. C content, 198
Transformation, of martensite, at constant temperature, 143, 156
    "split" or doubled, 157
Transverse properties, 93
Troostite, definition, 154
  etching characteristics, 193
  hardness of, 156
  primary, 133, 153
    formation, 156
    pearlitic structure, 156
  secondary, 186
Tungsten in precipitation hardenable alloys, 215
Turbine forgings, heat treatment, 140

U

Understressing in fatigue, 34

V

Valve steel, Cr-Ni-Al, 215
Vanadium in precipitation hardenable alloys, 216
Vickers pyramid hardness test, 11

W

Warping in quenching, 178
Water gas reaction, 311
Wear, resistance, 57
  testing, 57
Weldability, 56
Welding, hardening in, 142
Wire, patenting of, 142

X

X-magnification symbol, 78

Y

Yield point, 24
Yield strength, 24, Fig. 10, 22

[illegible]

- Neatsfoot oil - obtained by boiling
[illegible] bones of the feet
of horned cattle with

[illegible] oil

[illegible]

[illegible]

Light yellow color.

[illegible]

[illegible]

Flash - 440°F

[illegible]